

# TRANSACTIONS

*of the  
American Society  
for Steel Treating*

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# AGATHON ALLOY STEELS



*TRANSACTIONS*  
*American Society for Steel Treating*

VOL. XVI

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NO. 1

ON OXYGEN DISSOLVED IN STEEL, AND ITS  
INFLUENCE ON THE STRUCTURE

BY MARCUS A. GROSSMANN

*Abstract*

*It is shown that in regular pack carburizing (box carburizing), steel absorbs oxygen as well as carbon.*

*The amount of soluble oxygen present in steel appears to affect the microstructure. (The soluble oxygen diffuses readily, and is considered here apart from precipitated oxides and nonmetallic inclusions.) A high oxygen content seemingly favors solubility of cementite in alpha iron, and as a result of this solubility it affords opportunity for cementite to diffuse.*

*Cementite diffusion (and hence agglomeration) seemingly accounts for the phenomenon of "split cementite" in the structures observed in a McQuaid-Ehn test, and also for the structure of "boundary cementite" in low-carbon steels. These structures are characteristic of high-oxygen material.*

*Solubility of cementite in alpha iron can be indicated on the iron-carbon diagram. Since some steels exhibit cementite diffusion while others do not, it is suggested that different iron-carbon diagrams can be drawn for different steels, varying the diagram to accord with the cementite solubility and hence with the diffusion-power. Oxygen contents must apparently be considered, and in carburized steels it is the oxygen content after carburizing that is significant.*

**I**T WAS realized many years ago that oxygen might be present in solid steel, and it was naturally supposed that its presence might have an effect on the properties of the steel. A method

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for determining the amount of oxygen was developed by Ledebur (1, 2),<sup>1</sup> which involved heating the sample in a current of hydrogen and weighing the water which resulted from the reaction of the applied hydrogen with the oxygen in the steel. The method, with refinements added, is still in use today. There has been much exploratory work in the intervening years, attended with moderate success. More recently, however, a very fine technique has been developed for determination of the total oxygen. This technique has been developed at the Bureau of Standards in Washington, under the direction of L. Jordan, and at the Technische Hochschule in Aachen, Germany, under the direction of the late Prof. P. Oberhoffer and his associate W. Hessenbruch. The sample is melted in a carbon crucible in vacuum (with high frequency induction heating), and the carbon monoxide resulting from reaction of carbon with the oxygen is drawn off and determined. The method (3, 4) has given highly satisfactory results, and has been used throughout the present work. Indeed, it must be said that the present work has been possible only because the refinements in technique at the above two institutions have at last given an accurate and reliable method for oxygen determination. The great credit for first adopting high frequency melting in a graphite crucible is to be accorded to Jordan, in Washington. It is the basis of the present method. Great credit is also to be accorded to Hessenbruch, in Aachen, whose painstaking work in personally subjecting all existing procedures to critical examination, has given fundamental data on the accuracy and the limitations of the numerous methods, including the present one.

When oxygen is present in steel, there are obviously the possibilities that it occurs as oxides (nonmetallic inclusions or submicroscopic oxides) or that it is present as dissolved oxygen. Refinements in microscope technique have given valuable information on the distribution of oxides, silicates and other oxide compounds in ingots and in resulting steel articles. There are also indications that steel may at times contain submicroscopically precipitated oxides. In both cases (visible inclusions and submicroscopic particles), the oxygen would appear in the values found for total oxygen. These two forms are however not considered in the present paper. Only soluble oxygen will be considered. Evidence is offered that there is quite often a significant amount of oxygen in

<sup>1</sup>The figures in parentheses refer to the bibliography appended to the paper.

the dissolved form, that it diffuses readily at steel-treating temperatures, and that it has a decided and frequently encountered effect on the structure of the carbides.

### OXYGEN DIFFUSION

It seems important to demonstrate first that oxygen does actually diffuse in steel. It is readily imaginable that "dissolved" oxygen might diffuse in steel, whereas solid oxides would presumably diffuse with more difficulty. Mahin and Dillon (5), it is true, have offered microscopic evidence indicating that blast furnace slag, open-hearth slag and mill scale show diffusion phenomena into steel when heated in contact with it at 1650 degrees Fahr. (900 degrees Cent.), for from 10 to 20 hours. The evidence is most interesting as indicating some kind of change due to diffusion. Further, Harder, Weber and Jerabek (6) heated low-carbon steel in contact (a.) with magnetic mill scale ( $\text{Fe}_3\text{O}_4$ ), (b.) with ferrie oxide ( $\text{Fe}_2\text{O}_3$ ), (c.) with open-hearth slag from open-hearth iron, the slag presumably therefore being high in iron oxide, (d.) with aluminum oxide ( $\text{Al}_2\text{O}_3$ ) and (e.) with alundum (granular  $\text{Al}_2\text{O}_3$ ). In each case, some kind of diffusion effect was noticed, which caused "abnormality" in the affected zone after carburizing. They also observed what appeared to be iron oxide in the grain boundaries of steel heated 35 hours in carbon dioxide at 1725 degrees Fahr.

So far as the author is aware, however, evidence has been lacking that oxygen is quite soluble in steel in a form not seen under the microscope, and that it diffuses readily at normal steel treating temperatures.

Evidence of oxygen diffusion was found in carburizing, in an atmosphere of carbon monoxide (CO) gas.

### OXYGEN IN CARBURIZATION

The first tests were carried out on electrolytic iron, kindly furnished by Dr. Frederick C. Langenberg. A piece of the electrolytic iron was machined to about 2 inches square and about  $\frac{5}{32}$  inch thick, and was then cut into 6 slices, as illustrated in Fig. 1. Pieces 1, 3 and 5 were grouped to form one sample, and the oxygen content was determined. Then pieces 2, 4 and 6 were pack carburized at 1690 degrees Fahr. (920 degrees Cent.) for 8 hours, in a com-



mercial carburizing compound, resulting in carbon penetration to a depth of about  $\frac{1}{32}$  inch. These three carburized pieces were cleaned of scale, were then combined to form one sample, and the oxygen content determined. (A small piece was reserved from one of them for microscopic examination.) A comparison of the

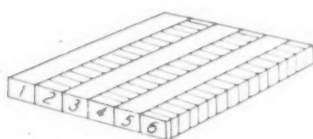


Fig. 1 — Diagram Showing Method of Cutting Samples for Carburizing Tests.

oxygen content before carburizing with the average content after carburizing showed the following:

Before carburizing . . . . .	0.0275 per cent oxygen
After carburizing . . . . .	0.0344 per cent oxygen

The figures thus show that a significant amount of oxygen has been absorbed by the iron during the carburizing. It should be mentioned at once that subsequent tests showed even much larger absorptions of oxygen.

Tests were made in the same way on certain open-hearth irons.

Table I  
Compositions of Irons

Material	A	B	C
Carbon . . . . .	0.04	0.03	(0.02)
Manganese . . . . .	0.12	0.05	0.02
Silicon . . . . .	0.009	0.003	(<0.01)
Copper . . . . .	0.25	0.04	0.05
Oxygen			
Before carburizing . . . . .	0.047	{ 0.036 0.033	0.031
After carburizing . . . . .	0.081	{ 0.078 0.077	0.059

The samples were prepared in the same manner as above (Fig. 1) so that the figures for each material are comparative within themselves but not necessarily with another material. The data are given in Table I. The tests were made in commercial carburizer, of the ordinary carbon variety for packing, but of unknown com-

position. It is seen that the oxygen absorption is very great indeed. Similar results have since been obtained when using a packing mixture of 60 per cent (by weight) charcoal and 40 per cent barium carbonate. The oxygen absorption is of the same order of magnitude as that reported here.

The figures of course give only the total amount of oxygen absorbed, i. e., the average for the whole piece, and give no indication of its concentration at different depths.

The significant conclusion is thus reached that in carburizing with CO gas, or at least with carburizing mixtures which when heated give an atmosphere of CO, carbon is not absorbed alone, but is accompanied by a certain proportion of oxygen. The amount of oxygen absorbed may equal or exceed the amount originally present in the steel. What effect this may have on the steel is discussed later on.

Reference must be made to extensive work by Schenck (7), who has studied the equilibrium of iron with CO and CO<sub>2</sub> at different temperatures. A very painstaking series of experiments led him to the conclusion that extraordinarily large amounts of oxygen are absorbed by iron at temperatures in the range studied here. His work leads him to postulate an "oxo-ferrit," or solid solution of oxygen in iron, containing variously from 2 per cent to 5 per cent oxygen. The amount appears quite high, and is altogether out of the range of anything found in the present experiments. Attention should be called to certain circumstances in his experiments which should be considered. The percentages of oxygen in the iron were not results of analysis, but were calculations from the equilibria with the measured amounts of CO and CO<sub>2</sub> in the reactions. Direct determinations of the amount of oxygen in the iron would be of interest. Also, it may be significant that he used iron in the form of powder. It has been found both at the Bureau of Standards and at the Technische Hochschule that the fineness of the sample used was of importance in the determination of oxygen. Hessenbruch (4) reports that in examining a single piece of steel, and taking samples of different thicknesses, the oxygen content increased regularly from 0.004 per cent when the sample was a single thick piece, to 0.082 per cent in the sample which was in the form of fine shavings. Jordan and Eckman (3) had also previously reported finding 0.020 per cent oxygen in a single solid sample, and

0.041 per cent when the same material was in the form of fine millings. The possibility of surface adsorption or surface oxidation from the air must apparently be considered. The adsorption (if that is the explanation) has taken place at room temperature or at least below 400 degrees Fahr. (200 degrees Cent.), because the chips and turnings showed no temper colors and must therefore have remained below 400 degrees Fahr. in the machining operation. If such disproportionate increases in oxygen can occur in fine turnings at low temperatures, a similar phenomenon may need to be examined in powder at carburizing temperatures.

In connection with oxygen absorption in carburizing, attention should also be called to the work of Guthrie and Wozasek (8), who showed that, in gas carburizing, the presence of oxygen in some form was at times advantageous.

#### STRUCTURE OF CARBURIZED LAYERS

One obvious application of the present discovery lies in considerations of the McQuaid-Ehn carburizing test. McQuaid and Ehn (9, 10) showed that certain irregularities sometimes occurred when heats of steel of like chemical composition were carburized and hardened, and that these irregularities bore some relation to differences observed in the microstructure when the carburized layer was examined before hardening. They astutely attributed the observed differences to variations in oxygen, either in the dissolved form or as precipitated oxides. The phenomenon has been studied many times in the light of their conjecture, especially at the Bureau of Standards (11) where oxygen analyses were made with the Bureau's apparatus for determining oxygen. The oxygen content was determined on steels which were then carburized and examined microscopically. No connection was observed between the total oxygen content of the steels and their degree of "normality" or "abnormality" in the McQuaid-Ehn test. Gat (12) has stated that high oxygen steels are abnormal, in the sense that they show "split cementite," and this seems in the main to be correct. But the inference that steels initially low in oxygen are normal must obviously be studied on a different basis. If indeed the structure bears some relation to the oxygen content, then the phenomena must be examined, not in relation to the original oxygen, but in relation to the original oxygen plus the oxygen absorbed in car-



burizing, or briefly the oxygen present in the carburized layer after carburizing.

Attention will now be called to certain structures observed in such carburized layers. It will appear that these structures represent one phase of a general situation, in which the distribution of carbides is affected by the presence of dissolved oxygen.

An extreme case of such an effect is found in electrolytic iron. It has been observed frequently that electrolytic iron, when carburized, is "highly abnormal," regardless of its original oxygen content. The expression "highly abnormal" means that on the outside of the specimen, in what would commonly be the hypereutectoid zone, there are found only large masses of cementite, lying in a matrix of ferrite, pearlite being absent, or present in only small quantities.

Fig. 2 is typical of the outside layer or hypereutectoid zone of electrolytic iron carburized and then cooled slowly. The structure is seen to consist of relatively large masses of cementite, associated with grains of ferrite. The striking feature is the absence of pearlite in this hypereutectoid zone. The carbon content is perhaps 1.25 per cent, and one would ordinarily expect pearlite grains surrounded by the hypereutectoid cementite. Instead there are only grains of ferrite, associated with the masses of cementite.

A structure similar to this was shown by Portevin (13) in discussing the paper by Ehn (10). His illustration is reproduced in Fig. 3. It represents Bessemer steel which was overblown so that "the degree of oxidation was excessive." The solid steel was then carburized and Fig. 3 represents the hypereutectoid zone. It is seen to consist of cementite and ferrite.

Attention may now be directed to another case in which massive cementite (instead of pearlite) is associated with grains of ferrite.

#### CEMENTITE IN LOW-CARBON, HIGH-OXYGEN STEELS

In some low-carbon steels, cementite occurs as "boundary cementite,"—that is to say, instead of being present in the form of pearlite patches, it is found agglomerated along the edges of ferrite grains. This phenomenon has been illustrated repeatedly in the literature (14, 15, 16), and is shown also in Fig. 4. The micrograph represents a commercial low-carbon open-hearth iron, with

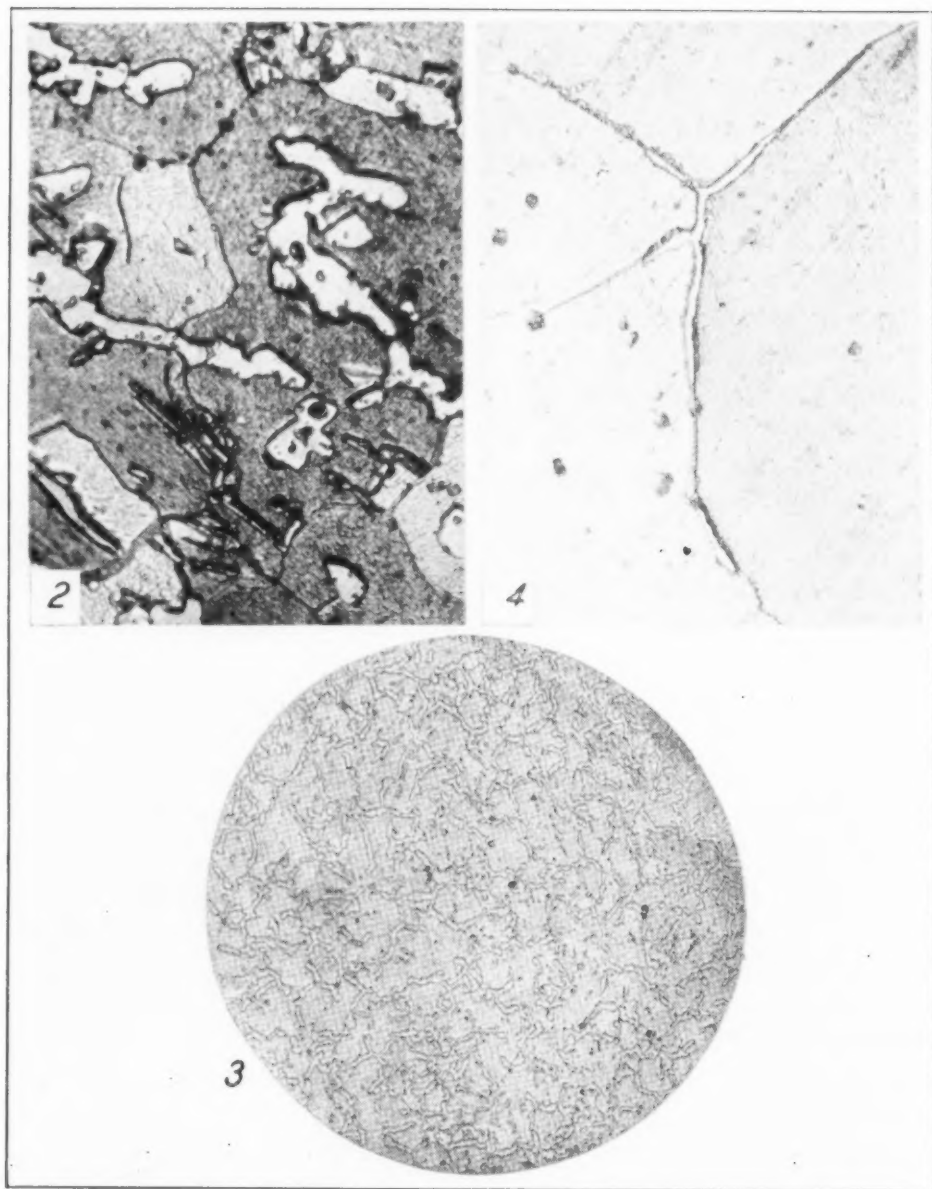


Fig. 2—Photomicrograph of "Highly Abnormal" Case on Carburized Electrolytic Iron.  $\times 1000$ . Fig. 3—Photomicrograph of Portevin's Structure of Excessively Oxidized Bessemer Steel, After Carburizing.  $\times 200$ . Fig. 4—Photomicrograph of Cementite in Open-Hearth Iron with 0.02 Per Cent Carbon; 0.02 Per Cent Manganese; 0.002 Per Cent Silicon; and 0.073 Per Cent Oxygen.  $\times 1000$ .

0.02 per cent carbon, 0.02 per cent manganese, 0.002 per cent silicon, and 0.073 per cent oxygen.

The fact that the cementite is in the grain boundaries has considerable significance. A restatement of certain elementary

facts will emphasize the point. When low carbon steel is in the austenitic condition (with the carbide all in solution) and is then slowly cooled, ferrite separates from the austenite during the critical range, the carbide (or carbon) becoming concentrated in the remaining austenite. At the  $A_1$  temperature, the remaining austenite will have reached the eutectoid composition, and on cooling through this temperature the austenite will transform to pearlite. It is generally assumed that the carbide was insoluble in the ferrite which precipitated during the early cooling, and that all of the carbide will be found in the pearlite. The pearlite consists of an intimate mixture of ferrite and carbide under these circumstances, but under these circumstances only. This type of structure is illustrated in Figs. 14 and 15. Fig. 14 represents the following composition: carbon 0.28 per cent; manganese 0.51 per cent; silicon 0.18 per cent; chromium 0.98 per cent; molybdenum 0.34 per cent; and oxygen 0.007 per cent. Fig. 15 represents a steel with carbon 0.30 per cent; manganese 0.77 per cent; silicon 0.17 per cent; chromium 0.90 per cent; molybdenum 0.14 per cent; and oxygen 0.004 per cent. Both samples were furnace-cooled from 1600 degrees Fahr. (870 degrees Cent.), taking 4 hours to cool to 1000 degrees Fahr. (540 degrees Cent.). The steel of Fig. 15 happens to be banded.

In the material shown in Fig. 4, the carbide is not present as pearlite patches, but as massive cementite, and it must therefore be assumed that some action is involved which is not included in the orthodox mechanism of the preceding paragraph.

It is perhaps advisable to introduce some evidence which indicates that the intergranular constituent is actually cementite. There was at first considerable doubt on this point in the mind of the author, and the question had been raised in the literature, since there was no accepted basis for assuming that the cementite could separate as shown. But the evidence all points to its being cementite. When etched with nitric acid in alcohol, it remains bright and sharply outlined as in Fig. 4. It is hard in a scratch test with a needle. It is brittle, as shown by the cracks in it in Fig. 5. The brittle break was observed when the polished piece showing such intergranular constituent was deformed severely and observed again on the polished face. The constituent breaks sharply into several pieces, and shows no signs of plasticity. The breaks are in-



dicated by arrows. The hardness, the brittleness and the fact that it is unresolved under the microscope all point to its being a compound. The thought that it might be a compound other than cementite led to etchings with numerous reagents, with the following results.

1. Neutral sodium picrate.....No attack
2. Hydrogen peroxide and sodium hydroxide.....No attack
3. Hot alkaline potassium ferricyanide  
10 grams  $K_3Fe(CN)_6$  + 10 grams NaOH in 100 cubic centimeters water .....Etched (1 minute)
4. Hot alkaline potassium permanganate.....Etched (1 minute)
5. Hot alkaline sodium picrate.....Etched (1 minute)
6. Hot 25 per cent sodium hydroxide.....Etched (5 minutes)
7. Boiling potassium ferricyanide with sodium hydroxide  
10 grams  $K_3Fe(CN)_6$  + 0.2 grams NaOH in 100 cubic centimeters water .....No attack  
10 grams  $K_3Fe(CN)_6$  + 0.8 grams NaOH in 100 cubic centimeters water .....Etched (10 minutes)
8. Copper sulphate and hydrochloric acid in water.....No attack
9. Hot potassium bichromate in water.....No attack
10. Hot potassium bichromate with sodium hydroxide.....No attack
11. Chromic acid.....Etched strongly

The behavior is the behavior of cementite, with the possible exception of reagent 7, where the long boiling and the similarity to reagent 3 raise a question as to the evidence. It can be said safely that the constituent etches like cementite.

The evidence (see above) other than etching thus indicates that the material is a compound, and the etching itself indicates that that compound is none other than cementite. The only consideration against the probability of its being cementite is the manner of its distribution in the grain boundaries of alpha iron.

The fact that it is found between the ferrite grains would seem to indicate solubility in the ferrite, and a subsequent complete separation by diffusion. It is this evidence of solubility of cementite in ferrite which is now to be considered.

#### SOLUBILITY AND DIFFUSION OF CEMENTITE IN FERRITE

It had been observed repeatedly that this boundary cementite is a characteristic feature of low-carbon steels having a relatively high oxygen content, say over 0.03 per cent oxygen.

It will be recognized therefore, that there are two points of similarity between the carburized electrolytic iron described above and the steel just examined containing boundary cementite, the

similarity including also Portevin's over-oxidized Bessemer steel:

- (1) Massive cementite is associated with grains of ferrite.
- (2) The materials are high in oxygen, although in the case of the electrolytic iron the oxygen was acquired in the carburizing process.

In considering the above evidence, P. D. Merica (17) made the suggestion that the oxygen which is absorbed in carburizing might be considered as diffusing as  $\text{FeO}_n$ , or better as O, along with the carbon, and that "the solubility situation of iron carbide in alpha iron, and consequently the form and distribution of any constituent in it, may be changed by presence of small amounts of  $\text{Fe}(\text{O})_n$  in solid solution." All of the evidence secured so far is in harmony with this theory. It appears that a sufficient increase in the dissolved oxygen content of alpha iron will cause cementite to be appreciably soluble in it. The extent of the cementite solubility, though slight, is still adequate to permit considerable diffusion on cooling. The diffusion, which all takes place below the temperature of  $A_1$ , will permit a total diffusion of much larger quantities of cementite than are actually soluble in the ferrite at any one time. The action during diffusion is considered briefly.

The electrolytic iron examined above will furnish structures illustrating the mechanism of diffusion. A number of pieces of electrolytic iron were carburized to a hypereutectoid composition on the outside of the piece by heating for 12 hours in charcoal and barium carbonate, at 1690 degrees Fahr. (920 degrees Cent.). From this temperature the whole carburizing pot was cooled in the furnace to about 1420 degrees Fahr. (770 degrees Cent.), and a piece of the carburized electrolytic iron taken quickly from the pot and quenched in 10 per cent sodium hydroxide solution. The pieces were all about  $\frac{1}{8}$  inch thick, so that the speed of quenching was quite rapid. The structure is shown in Fig. 6, and is seen to consist entirely of martensite. The quenching has evidently been done from above  $A_{cm}$ , with the hypereutectoid cementite all in solution before the quench. The carburizing pot (with the remaining pieces in it) was then cooled in the furnace to about 1300 degrees Fahr. (700 degrees Cent.), and a piece withdrawn from the pot and quenched as before. The structure is shown in Fig. 7. Evidently the quenching has been done from just above  $A_1$ , since the hypereutectoid cementite has separated around the austenite grains. This is the normal expected behavior, and in

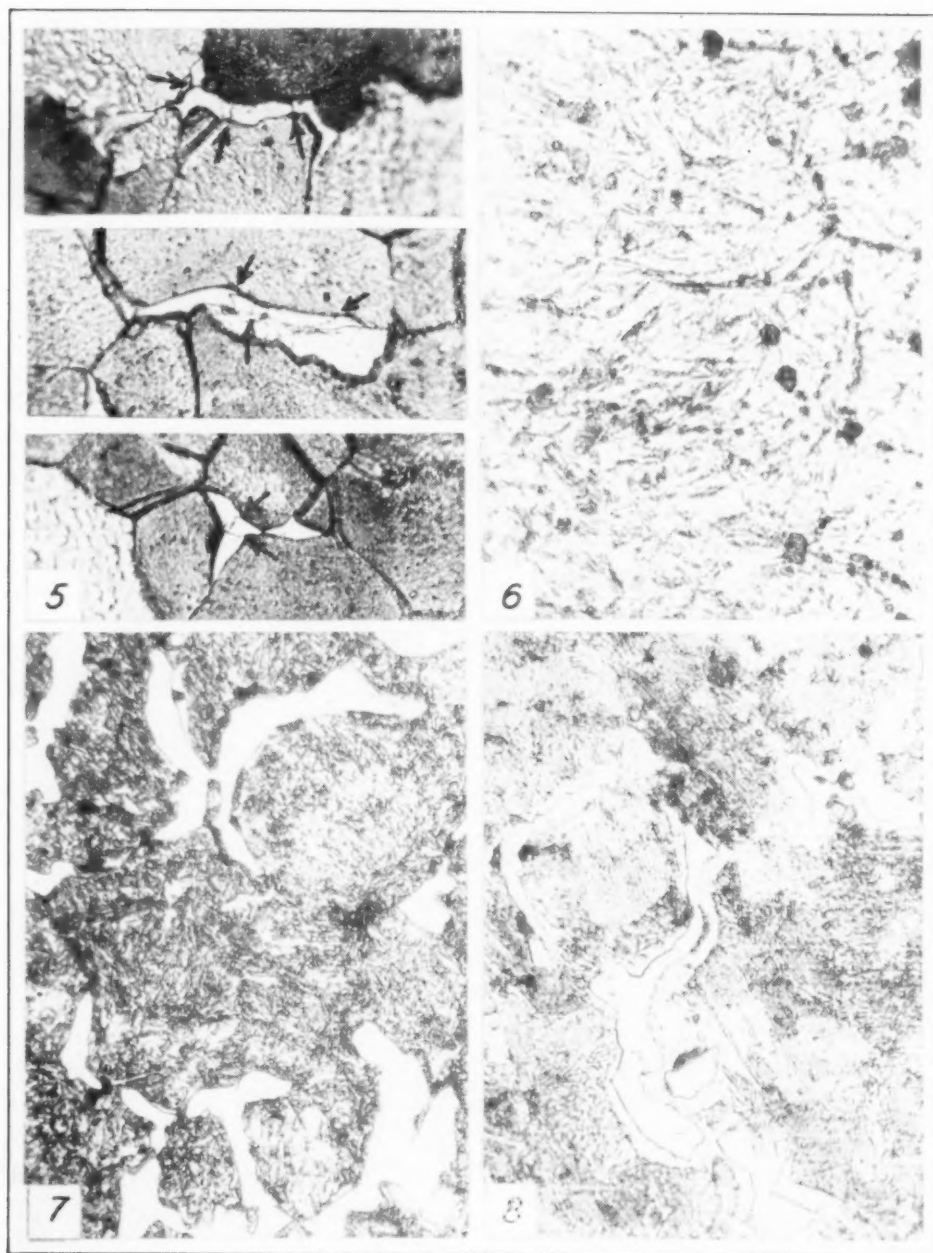


Fig. 5—Photomicrograph of Brittle Fracture of Cementite. Breaks Indicated by Arrows.  $\times 1000$ . Fig. 6—Photomicrograph of Carburized Electrolytic Iron, Hypereutectoid Zone, Quenched from 1420 Degrees Fahr., just Above Acm.  $\times 1000$ . Fig. 7—Photomicrograph of Same Iron as Fig. 6, but Quenched from 1300 Degrees Fahr., just Above A<sub>1</sub>.  $\times 1000$ . Fig. 8—Photomicrograph of Same Iron as Fig. 6, but Air-Cooled from 1300 Degrees Fahr.  $\times 1000$ .

this treatment carburized electrolytic iron has exhibited no unusual features.



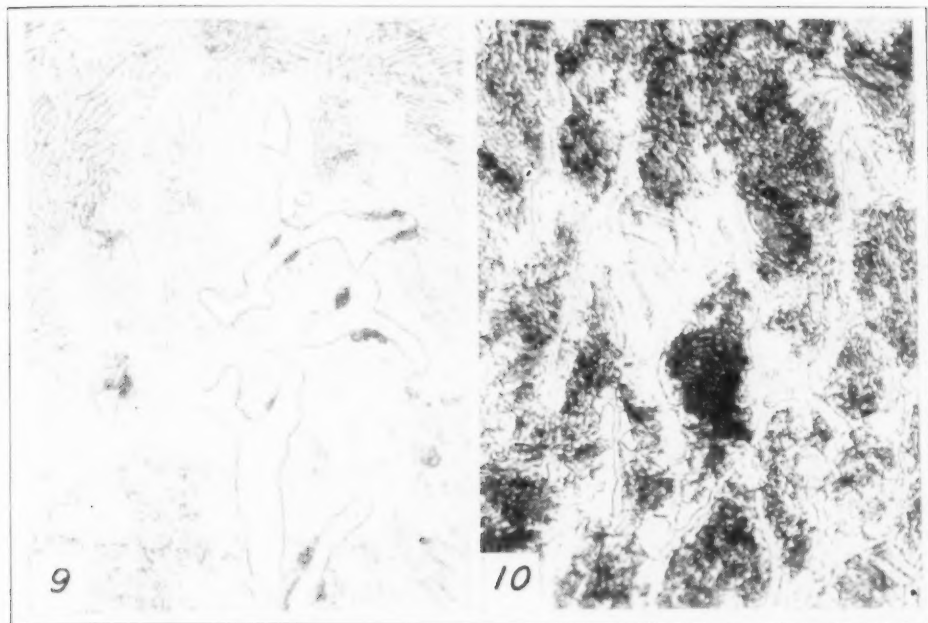


Fig. 9—Photomicrograph of Same Iron as Fig. 6, but Pot-Cooled from 1300 Degrees Fahr.  $\times 1000$ . Fig. 10—Photomicrograph of Same Iron as Fig. 2, but Air-Cooled from Carburizing.  $\times 1000$ .

If, however, the carburized electrolytic iron is taken from the pot at 1300 degrees Fahr. (700 degrees Cent.) and cooled in air, a new effect will be observed. The structure is illustrated in Fig. 8. The hypereutectoid cementite is found as before surrounding the original austenite grains. The major portion of the original austenite grain consists of pearlite. But between the pearlite and the cementite is a layer of ferrite. That this is ferrite is indicated by a number of circumstances: It is soft, it merges directly into the ferrite of the pearlite, when etched more deeply it has the texture characteristic of ferrite (or more particularly of high-oxygen ferrite), and when polished and deformed it gives the wavy slip lines commonly found in ferrite.

The reason for the existence of this band of ferrite is seemingly as follows: it was originally a region of pearlite, but the cementite in it has migrated out to build up on the hypereutectoid cementite. The tendency of the cementite and the ferrite is to separate as two phases, each phase agglomerated as much as possible. The tendency to separate is evidenced in the pearlite by the separation of the cementite from the ferrite in individual lamellae. Existence in the form of lamellar pearlite, however, is evidence

that there has been little subsequent opportunity for agglomeration of the cementite, or of the ferrite. The two phases are apparently so insoluble in each other that there has been either no opportunity or insufficient time for diffusion. When the austenite has transformed, the individual masses or crystallites of ferrite and cementite must remain in the places they occupy upon transformation. It is such a structure of pearlite which is illustrated in Figs. 14 and 15. But the presence of a certain amount of oxygen in solution in the ferrite seemingly causes the cementite to be slightly soluble in it. This slight solubility permits the cementite which was in the pearlite to diffuse out from the pearlite and crystallize upon the cementite in the grain boundaries, and form a crystalline entity with it.

This is the structure illustrated in Fig. 8. In the outer portions of the original austenite grain, the cementite which was in the pearlite in that region has migrated out and joined the hyper-eutectoid cementite, leaving a band of ferrite between the central pearlite area and the bounding cementite plates.

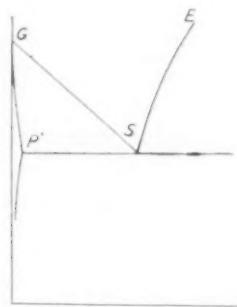


Fig. 11 — Iron-Carbon Diagram Indicating Some Solubility of Cementite in Alpha Iron.

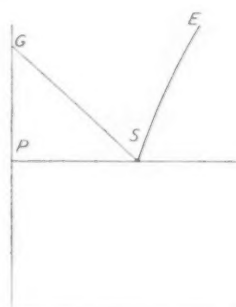


Fig. 12 — "Ordinary" Iron - Carbon Diagram, Cementite Insoluble in Alpha Iron.

Now if the cooling is still slower, then more of the cementite in the pearlite will have time to diffuse outward and join the cementite in the bounding regions. Such a condition is illustrated in Fig. 9. A sample of the same electrolytic iron was left in the same carburizing pot when used for Fig. 8 and the whole cooled in mica from 1300 degrees Fahr. (700 degrees Cent.). It will be seen that the ferrite band is much wider, seemingly much more of the cementite having diffused out to join the boundary cementite. A still more pronounced case is the one illustrated in Fig. 2, where all of the cementite has migrated out to join the hypereutectoid

cementite, leaving only alpha iron and no pearlite. This is evidently also the case in Portevin's sample, shown in Fig. 3.

The sample of Fig. 2 was cooled in the pot. A more rapid cooling of the same iron as in Fig. 2 is shown in Fig. 10, which represents the structure after cooling in air. The separation is evidently not so complete.

#### THE IRON-CARBON DIAGRAM

Evidence is thus offered that under certain circumstances we find carbide diffusion below the temperature of  $A_1$ , and the diffusion would be explained readily by a small but appreciable solubility of cementite in alpha iron. This would result in an iron-carbon diagram of the type shown in Fig. 11. The difference between this and the "ordinary" iron-carbon diagram of Fig. 12 is in the solubility of cementite in alpha iron, which is indicated at the point  $P'$  in Fig. 11. A number of investigators (16, 18, 19, 20) have expressed their belief that cementite is actually soluble in alpha iron, the estimates of solubility varying from 0.03 to 0.05 per cent carbon, as carbide. The present experiments seem likewise to indicate that a solubility, of about the order of magnitude stated, exists under certain circumstances.

But it hardly seems probable that this applies to all steels. If cementite were soluble in alpha iron in all cases, then it would be reasonable to expect that the diffusion phenomena just noted would be common to all steels. It is well known that this is not the case. Thus Fig. 13 shows the hypereutectoid zone of a carburized piece of steel containing initially carbon 0.22 per cent, manganese 0.84 per cent, and silicon 0.07 per cent. This specimen was cooled slowly in the carburizing pot, requiring perhaps 45 minutes to cool from 1650 degrees Fahr. (900 degrees Cent.) to 1000 degrees Fahr. (540 degrees Cent.). It is seen that the pearlite grains are bounded sharply by the hypereutectoid cementite, and that there is no evidence of that type of diffusion below the temperature of  $A_1$  which is discussed earlier in this paper.

It appears therefore that in some steels there is solubility and diffusion of cementite in alpha iron, resulting in the structures of Figs. 2 to 10, whereas in other steels the lack of such solubility results in the structures of Figs. 13, 14 and 15.

It would seem justifiable to conclude that the reactions which produce the structure of Fig. 2 are represented by a diagram such

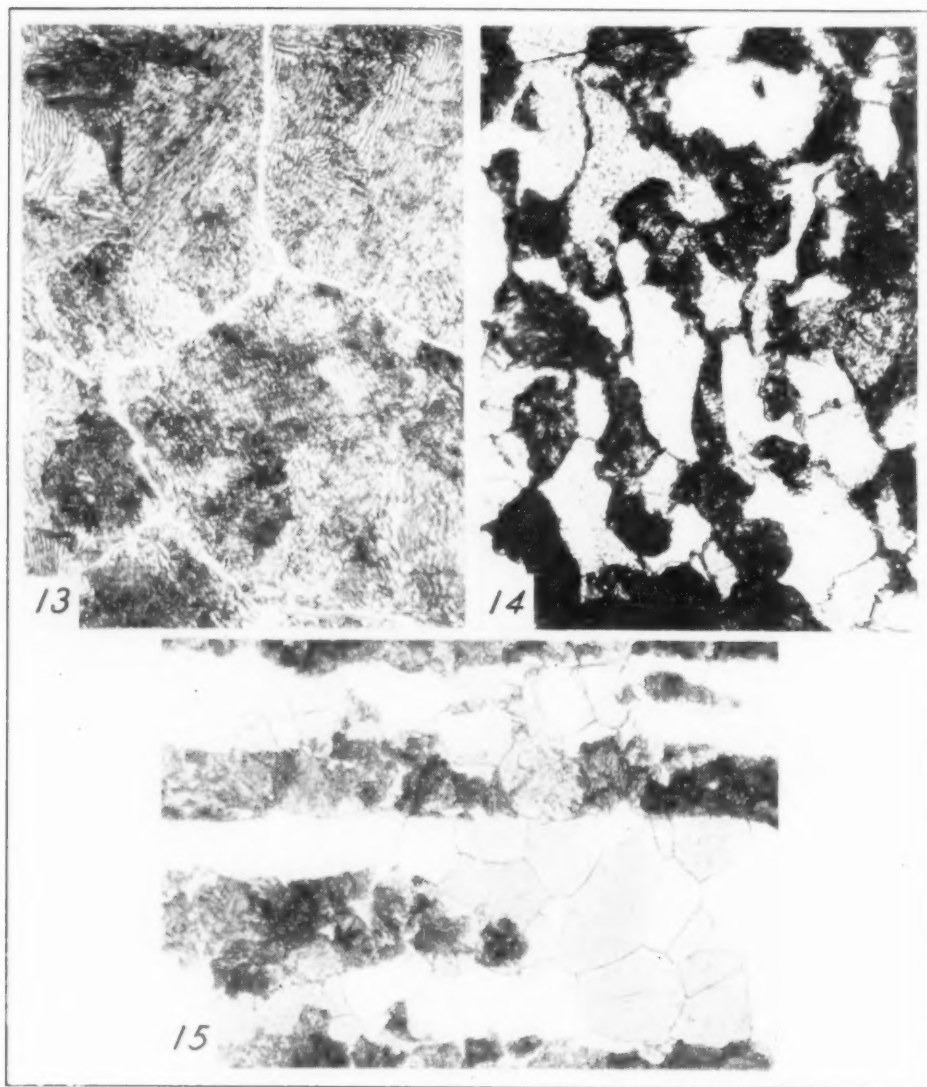


Fig. 13—Photomicrograph of Hypereutectoid Zone of Carburized Steel Containing Originally 0.22 Per Cent Carbon, and 0.84 Per Cent Manganese, and 0.07 Per Cent Silicon. Fig. 14—Photomicrograph of Pearlite in Steel Containing 0.28 Per Cent Carbon; 0.51 Per Cent Manganese; 0.18 Per Cent Silicon; 0.98 Per Cent Chromium; 0.34 Per Cent Molybdenum; 0.007 Per Cent Oxygen.  $\times 1000$ . Fig. 15—Photomicrograph of Pearlite in Steel Containing 0.30 Per Cent Carbon; 0.77 Per Cent Manganese; 0.17 Per Cent Silicon; 0.90 Per Cent Chromium; 0.14 Per Cent Molybdenum; 0.004 Per Cent Oxygen.  $\times 750$ .

as Fig. 11, whereas the reactions involved in the structure of Fig. 13 are represented by a diagram similar to Fig. 12.

The ferrite of plain carbon steels thus appears to be a complex and little-understood material. It is suggested that discrepancies found in the published data for cementite solubility in alpha iron may have to do with the oxygen content of the iron.

## McQUAID-EHN CARBURIZING TEST

The above data have a certain obvious bearing on the structures observed in the McQuaid-Ehn carburizing test. It is believed that diffusion phenomena can to advantage be studied in the consideration of "split cementite" (the band of ferrite between pearlite and cementite in the hypereutectoid zone of carburized steel).

It should be pointed out that no evidence has been adduced here to connect oxygen contents with grain size in the McQuaid-Ehn test. Grain size appears to be governed by laws which as yet have no evident connection with oxygen absorbed in carburizing.

It should also be pointed out that oxygen is not by any means the only element that may affect the solubility of carbide in alpha iron. It is conceivable that all of the ordinary steel-making elements, and also the alloys, affect that solubility, and that they may either lessen or increase the solubility. Indeed, evidence is as yet completely lacking as to the solubility or insolubility of cementite in ferrite free of manganese and silicon and also free of oxygen. Much further work will have to be done to determine the inter-relationship of oxygen and the common steel-making elements in cementite solubility and diffusion.

## SUMMARY AND CONCLUSIONS

1. It is shown that in ordinary pack-carburizing of low-carbon steel, carbon is not absorbed alone, but is accompanied by absorption of a certain amount of oxygen which diffuses into the steel.
2. In certain steels, cementite may occur not as pearlite but as massive cementite associated with grains of ferrite. The separation of pearlite into massive cementite and ferrite appears to be due to diffusion.
3. The phenomenon of "split cementite" in carburized steels appears to be due to this same diffusion.
4. The diffusion seems to be favored by high oxygen content in the steel, which points to oxygen as favoring solubility of cementite in ferrite.
5. McQuaid-Ehn phenomena should be examined in the light of original oxygen in the steel plus oxygen absorbed in carburizing.



6. If the solubility of cementite in alpha iron varies for different steels, these variations can be indicated by changes in the iron-carbon diagrams which represent those different steels.

#### ACKNOWLEDGMENTS

The author is greatly indebted to J. C. Joublane, who not only set up the oxygen-determining apparatus but also made all of the oxygen estimations reported in the present paper; to L. Jordan and W. Hessenbruch for many hints and valuable data in improved technique; and to C. C. Snyder for assistance in some of the microscopic work.

He would like also to acknowledge a personal indebtedness to two friends, Dr. Frederick C. Langenberg for constant encouragement and discussion and the furnishing of valuable samples, and Dr. Paul D. Merica for much generous counsel and for subtle and discerning consideration of the data.

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### DISCUSSION

**Written Discussion:** By S. Epstein and L. Jordan, Bureau of Standards, Washington, D. C.

Oxygen analyses by the vacuum fusion method at the Bureau of Standards<sup>1</sup> of normal and abnormal steel and ingot iron specimens have not confirmed the author's finding that an appreciable amount of oxygen is absorbed during carburizing. The accompanying table gives the results of these analyses. The specimens were similar in size to those used by the author and were carburized at 1725 degrees Fahr. (940 degrees Cent.) for 36 hours, and 9 hours, respectively. Neither for the longer nor the shorter period of carburization did the analyses indicate any significant increase in oxygen content. The ingot iron specimen which was carburized for 36 hours contained 0.049 per cent oxygen before carburizing and after the lengthy period of carburization the oxygen content decreased to 0.037 per cent. The ingot iron specimen which was carburized for 9 hours showed no difference in oxygen content before and after carburization. It should be noted that the accompanying table contains results of oxygen analyses of carburizing steels as well as ingot iron, whereas in the paper oxygen analyses of ingot iron only are given. While the results shown in the table do not disprove the possible occurrence of small changes in oxygen content during carburizing, they do not confirm the finding of the author that appreciable amounts of oxygen are absorbed during carburizing.

Irrespective of whether or not oxygen is absorbed during carburizing, the discussion given in the paper on the influence of oxygen on the structure of carburized steel is a stimulating contribution to our ideas on the subject. The view is suggested by the author that dissolved oxygen may make cementite more soluble in ferrite. One of the characteristic features of the structure of the carburized layer of abnormal steel is the presence of ferrite surrounding relatively large masses of cementite. The presence of this ferrite is explained by the rapid diffusion of the pearlitic cementite which takes place below the  $A_1$  point (owing to the assumed increased solubility of the cementite in the ferrite) and the consequent precipitation of the pearlitic cementite on the hypereutectoid cementite particles. Figs. 8 and 9 of the paper indicate that the formation of the ferrite layers about the cementite actually occurs below the  $A_1$  point. However, it is not clearly brought out that the other complementary feature of the abnormal structure, namely, the presence of large masses of coalesced cementite, cannot be due to diffusion below the  $A_1$  point. As shown in Fig. 7 of the paper the

<sup>1</sup>By H. C. Vacher, assistant scientist, and R. J. Kranauer, assistant scientific aid.

**Results of Oxygen Analyses by the Vacuum Fusion Method of Normal and Abnormal Steel and Ingot Iron Specimens, Before and After Carburizing**

Description of Specimen	Type	Condition	O <sub>2</sub> Wt. Per Cent
No. 113 Killed Steel	Abnormal	not carburized	nil
Al. added in mold			nil
0.14% C, 0.44% Mn		carburized	0.004
		36 hrs.—1725 degrees Fahr. (940 degrees Cent.)	0.003
No. 125 Effervescing Steel	Abnormal	not carburized	0.001
Al. added in mold			0.004
0.14% C, 0.42% Mn		carburized	0.002
		36 hrs.—940 degrees Cent.	0.003
No. 122 Killed Steel	Normal	not carburized	0.001
No mold addition			nil
0.14% C, 0.42% Mn		carburized	0.003
		36 hrs.—940 degrees Cent.	0.002
No. 134-0 Effervescing Steel	Normal	not carburized	nil
No mold addition			0.002
0.12% C, 0.43% Mn		carburized	0.005
		36 hrs.—940 degrees Cent.	
A Ingot Iron	Abnormal	not carburized	0.050
			0.048
		carburized	0.037
		36 hrs.—940 degrees Cent.	0.037
B Ingot Iron	Abnormal	not carburized	0.067
			0.068
		carburized	0.067
		9 hrs.—940 degrees Cent.	0.068
No. 134 I Effervescing Steel	Normal	not carburized	0.012
No mold addition			0.013
0.12% C, 0.43% Mn		carburized	0.010
		9 hrs.—940 degrees Cent.	0.010

coalescence of the cementite is already well marked above the  $A_1$  point. This feature of the structure, that is the coalescence of the cementite had been previously attributed to the lower solubility of cementite in abnormal austenite.<sup>2</sup>

To demonstrate whether or not the presence in the carburized layer of abnormal steel of coalesced cementite, (instead of thin envelopes of cementite in the grain boundaries as in normal steel) can be attributed to the lower solubility of cementite in abnormal austenite, the following series of heat treatments was made. Several pieces ( $1\frac{1}{2}'' \times 1'' \times \frac{1}{4}''$ ) of normal and abnormal steel and of ingot iron which was extremely abnormal, were carburized at 1725 degrees Fahr. (940 degrees Cent.) for 36 hours, so that a thick carburized layer was obtained. The carburized pieces were then heated to 1965 degrees Fahr. (1075 degrees Cent.) for 3 hours to insure the complete solution of the cementite, and air-cooled. The specimens so treated were used as working material for the following treatments, different specimens of each kind of steel being used for each treatment.

Heated at 1000 degrees Cent. for 3 hours, air-cooled  
Heated at 975 degrees Cent. for 3 hours, air-cooled

<sup>2</sup>O. E. Harder, L. J. Weber, and T. E. Jerabek. "Studies on Normal and Abnormal Carburizing Steel." TRANSACTIONS, American Society for Steel Treating, Vol. 13, 1928, pp. 961-1008.

O. E. Harder and W. S. Johnson. "Solubility of Carbon in Normal and Abnormal Steel." Preprint before 1928 American Society for Steel Treating Convention, Philadelphia.

Heated at 950 degrees Cent. for 4 hours, air-cooled  
Heated at 925 degrees Cent. for 4 hours, air-cooled  
Heated at 900 degrees Cent. for 4 hours, air-cooled  
Heated at 875 degrees Cent. for 6 hours, air-cooled  
Heated at 850 degrees Cent. for 6 hours, air-cooled  
Heated at 825 degrees Cent. for 6 hours, air-cooled

The specimens were packed in a carburizing compound to guard against oxidation and for each treatment the specimens were all heated together in the same pot under identical conditions.

Fig. 1 shows the structures of the normal and abnormal steel and ingot iron specimens after carburizing. The ingot iron specimen is very abnormal.

Fig. 2 shows the structures after heating to 1965 degrees Fahr. (1075 degrees Cent.) and air cooling. It will be noted that neither the abnormal steel nor the ingot iron exhibits coalescence of the cementite. In the abnormal specimens as well as in the normal specimen the cementite is present as needles along the cleavage planes and as a thin boundary network. It is thus apparent that upon heating to a high enough temperature to completely dissolve the cementite, no coalescence of the cementite occurs upon air cooling. It is of interest to note that in the abnormal steel the grain size is finer than in the normal steel or in the ingot iron.

Upon reheating to 1830 and 1790 degrees Fahr. (1000 and 975 degrees Cent.) respectively as shown in Figs. 3 and 4, the cementite again appears in the form of needles or plates and as a thin network around the grains.

However, upon heating at 1790 degrees Fahr. (975 degrees Cent.) as shown in Fig. 5, a feature appears which is of considerable interest, and which was noticeable to an even greater degree at the lower temperatures of reheating, namely, the concentration of masses of cementite near the surface. This took place to a larger extent in the abnormal specimens than in the normal specimens.

According to the iron-carbon equilibrium diagram a rather large amount (1.4%) of carbon can be held in solution at 1790 degrees Fahr. (975 degrees Cent.) still it does not appear that the high concentration of cementite at the surface shown in the photomicrographs at the left of Fig. 5, was at any time held totally in solution. As has been stated, at the lower heating temperatures the concentration of cementite at the surfaces was even more pronounced. The shape of the cementite masses at the surfaces is distinctly different from the cementite needles and boundary films adjacent to the surface. The appearance of the latter indicates it was held in solution, but that of the former would indicate that it was not in solution. The conclusion may be drawn that the concentration of cementite at the surface occurred not by total solution in the austenite but by diffusion of carbon in the austenite and its precipitation on preexisting nuclei of cementite or other particles. Precipitation might occur because of slight temperature fluctuations or because of local variations in chemical composition. For instance at a spot high in oxygen the solubility of cementite in austenite might be lowered and precipitation of cementite take place. Such diffusion in the austenite above  $A_1$  is analogous to diffusion in the

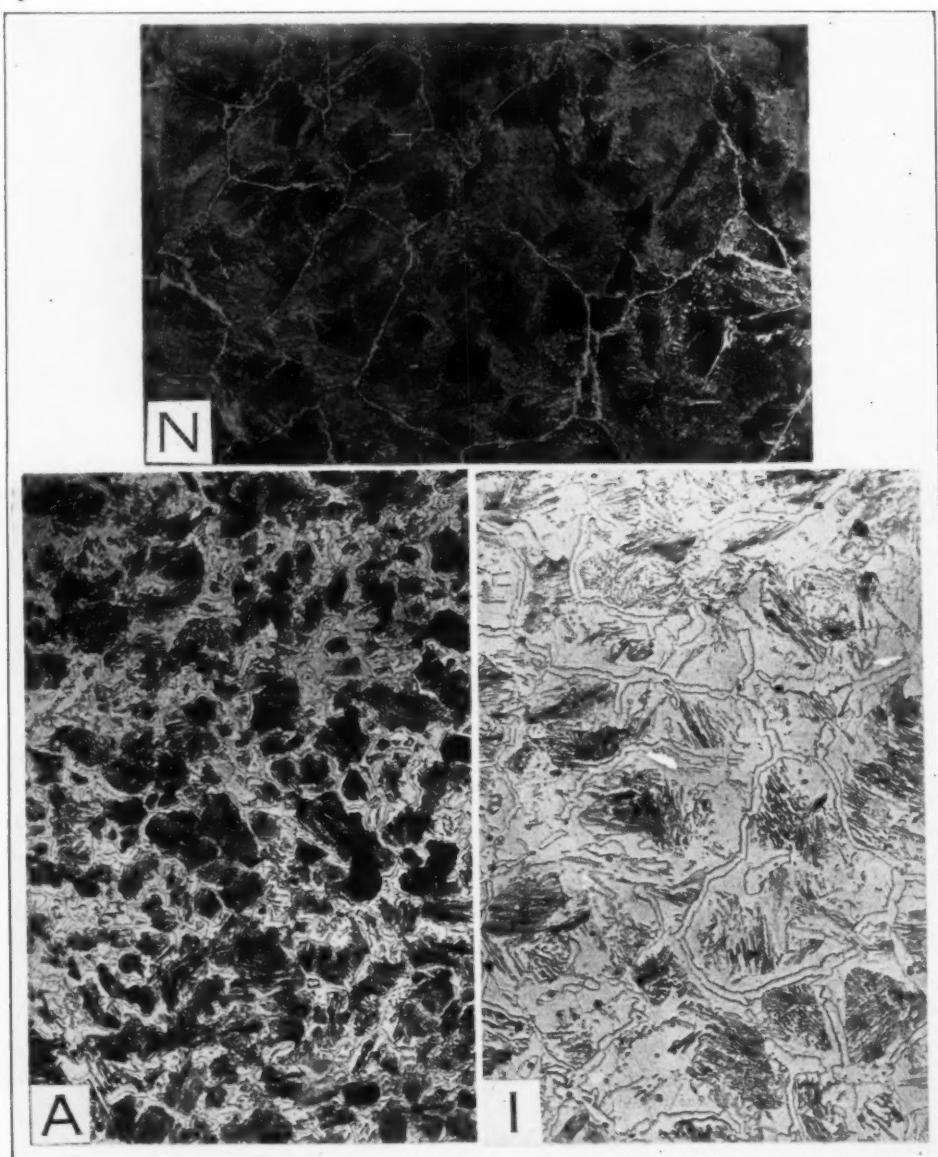


Fig. 1—Photomicrographs showing structures of the Carburized Layers of the Specimens Used. The Specimens were Carburized at 1725 degrees Fahr. (940 degrees Cent.) for 36 Hours in a Commercial Carburizer and Furnace-Cooled. N Indicates the Normal Specimen; A, the Abnormal specimen; and I, the Ingot Iron Specimen.  $\times 100$ . Etched with 2 per cent Alcoholic Nitric Acid.

As May be Seen the Ingot Iron Specimen was Very Abnormal in Structure.

ferrite below  $A_1$ , except for the greater solubility of cementite in austenite than in ferrite. It is probable that such diffusion of the cementite in austenite plays a part in the formation of the coalesced cementite in abnormal steel as well as in the absorption and penetration of carbon into the carburized layer during carburizing.



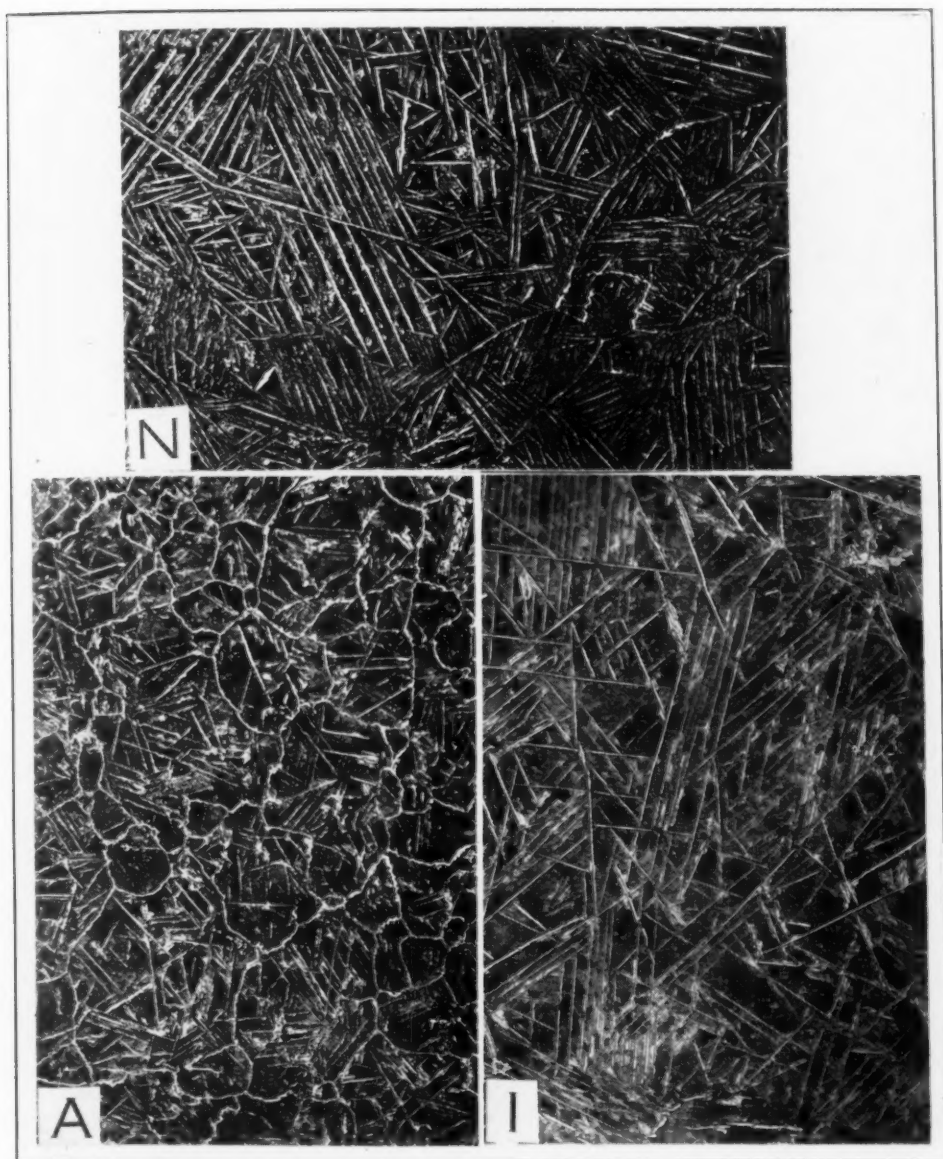


Fig. 2—Photomicrographs Showing the Structures After Heating at 1965 degrees Fahr. (1075 degrees Cent.) for 3 Hours, Air-Cooled. After heating at this Temperature no Coalesced Cementite Appeared in the Abnormal Specimens  $\times 100$ . The Specimens Shown in the Following Figures were Carburized, Heated at 1965 degrees Fahr. (1075 degrees Cent.) for 3 Hours and Air-Cooled, and then Reheated to the Temperatures Stated.

Fig. 6 shows the structures after heating at 1740 degrees Fahr. (950 degrees Cent.). In the normal steel the cementite is again present as plates and thin grain boundaries. The abnormal steel, however, shows signs of coalescence of the cementite. In Fig. 7 of the structures after heating at 1700 degrees Fahr. (925 degrees Cent.) coalesced cementite is present in all of the specimens, normal as well as abnormal. Figs. 8 to 11, of the struc-

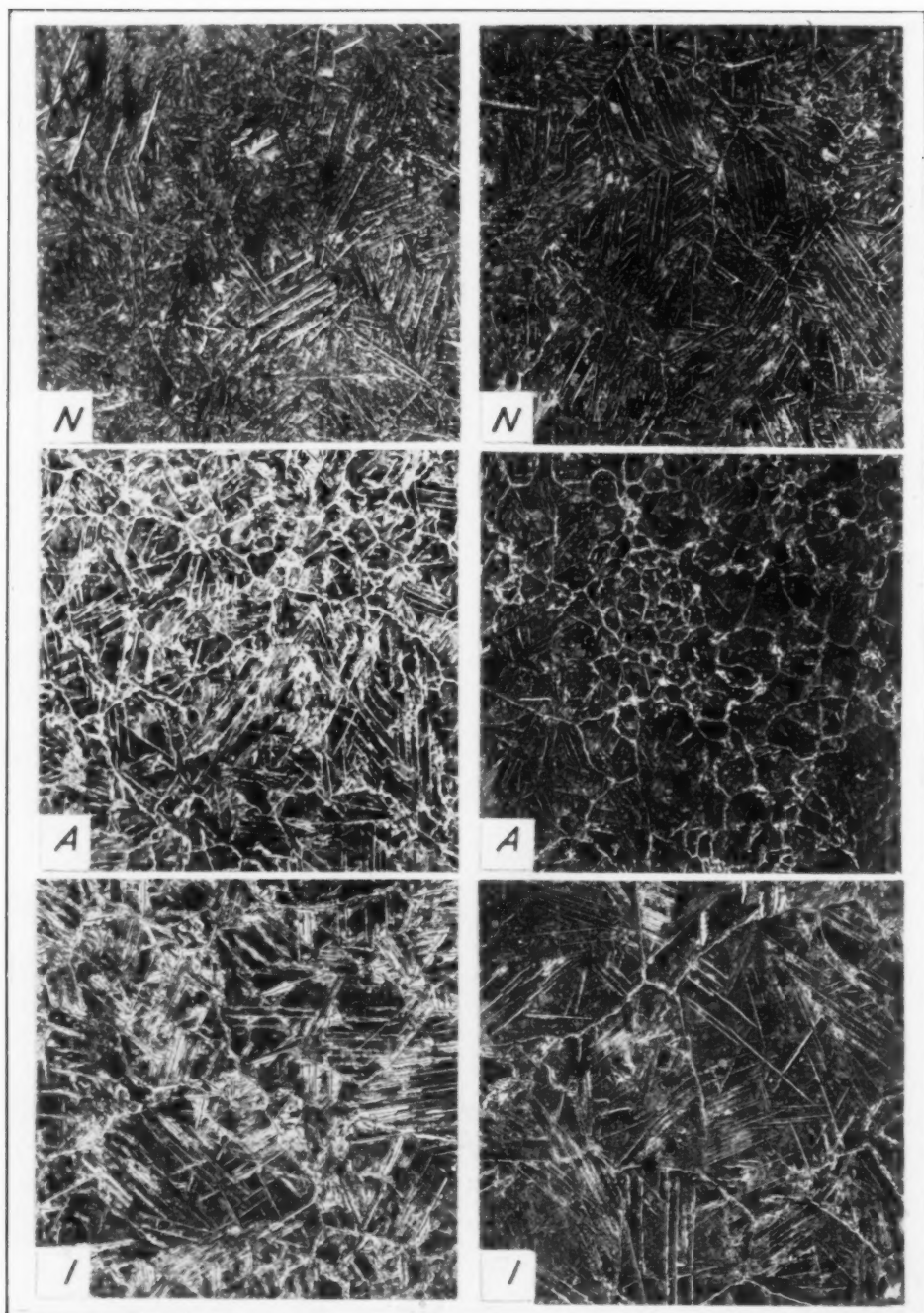


Fig. 3—Photomicrographs of Specimens Heated at 1830 degrees Fahr. (1000 degrees Cent.) for 3 Hours and Air-Cooled. (Left hand column)  $\times 100$ .

Fig. 4—Heated at 1785 degrees Fahr. (975 degrees Cent.) for 3 Hours and Air-Cooled. (Right hand column)  $\times 100$ .

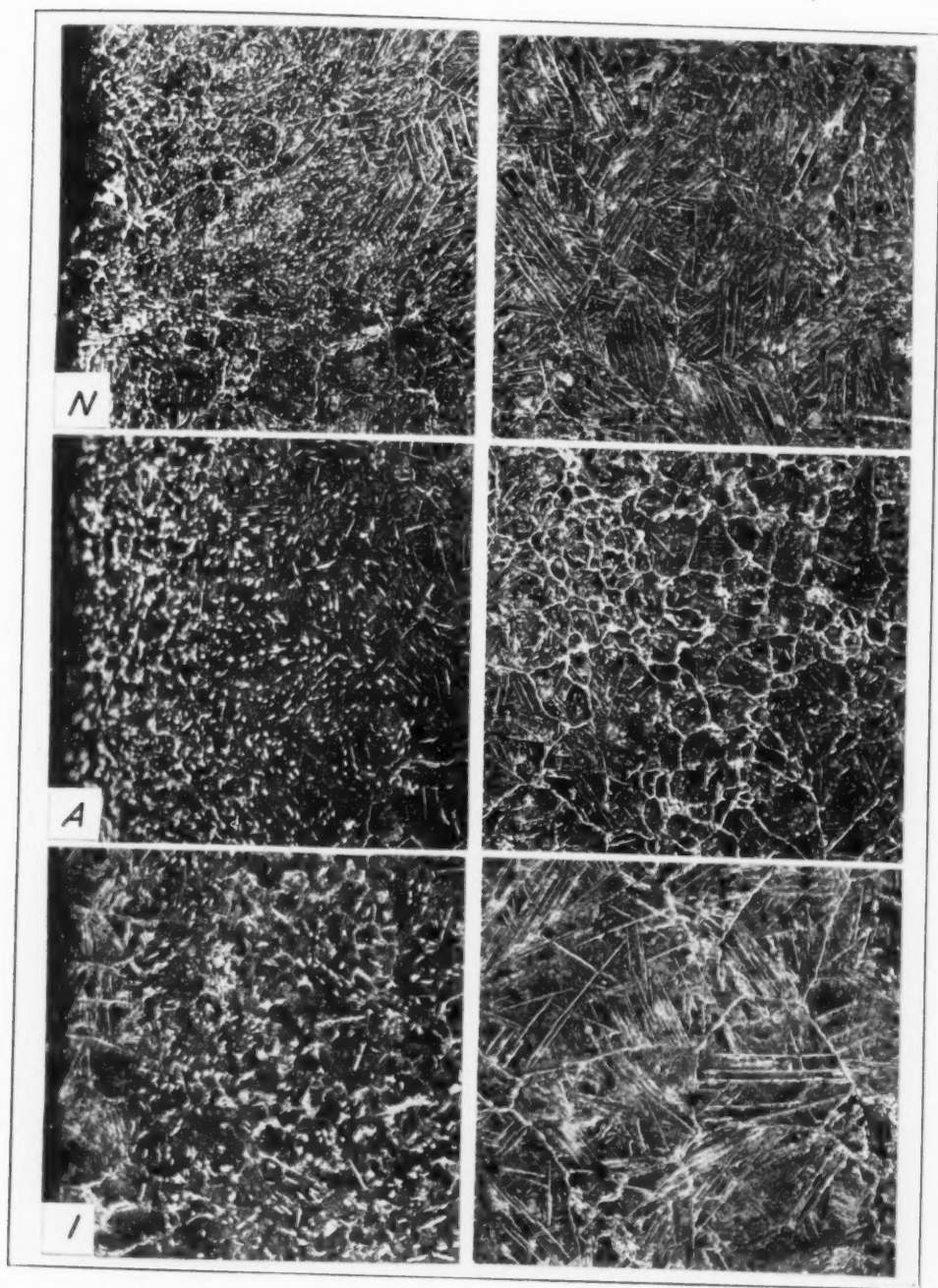


Fig. 5—Photomicrographs Showing Concentration of Cementite at the Surface. Heated at 1740 degrees Fahr. (950 degrees Cent.) for 3 Hours and Air-Cooled,  $\times 100$ . For each Specimen the Photomicrograph at the Left is at the Surface and the One at the Right is Immediately Adjacent to the Surface Toward the Interior. The Concentration of Cementite at the Surfaces was more Pronounced in the Abnormal Specimens.



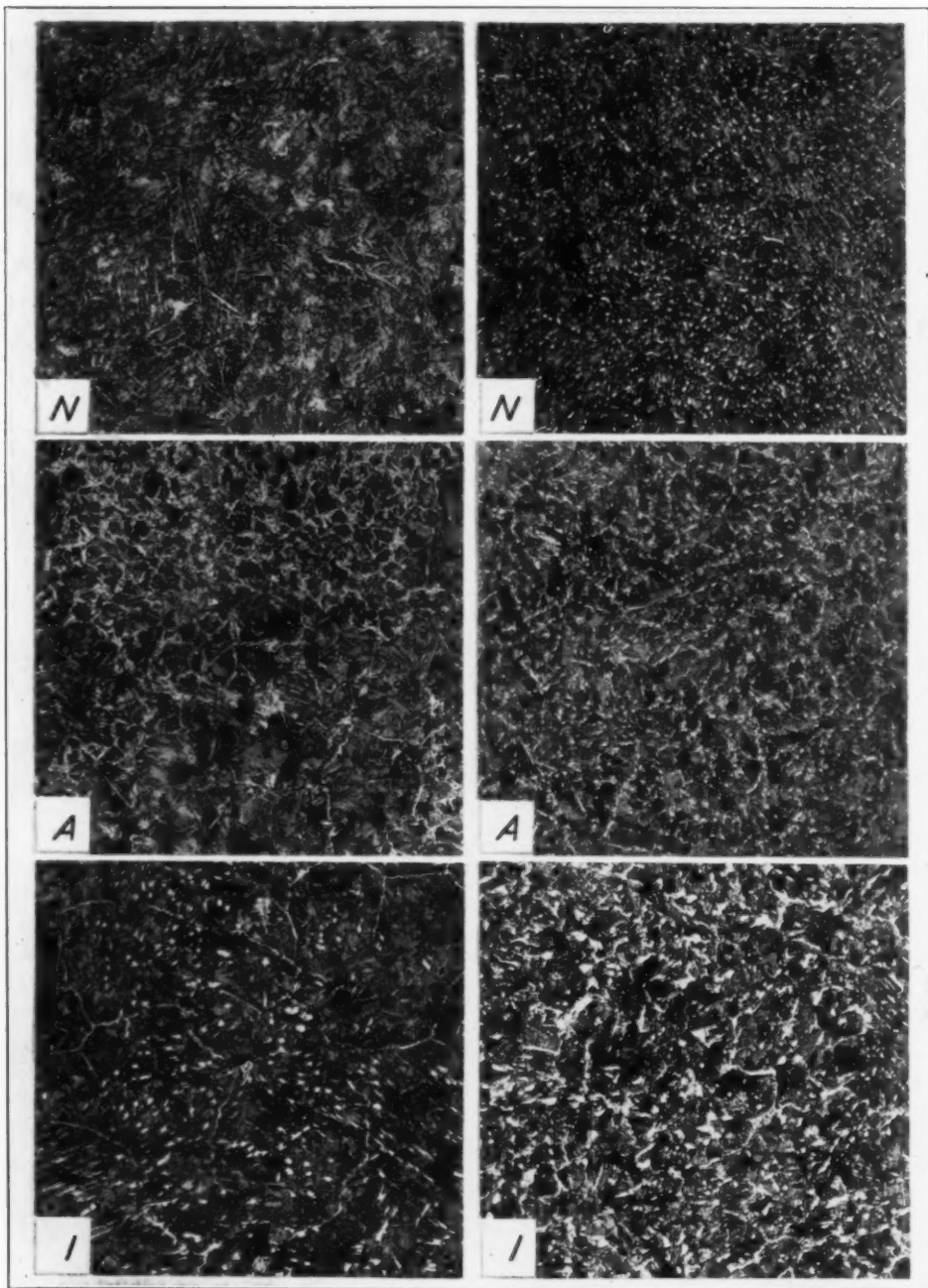


Fig. 6—Photomicrographs of Specimens Heated at 1740 degrees Fahr. (950 degrees Cent.) for 4 Hours and Air-Cooled. (Left hand column)  $\times 100$ .

There is No Coalescence of the Cementite in the Normal Specimen, but the Abnormal Specimens show Unmistakable Signs of Coalescence. Fig. 7—Heated at 1695 degrees Fahr. (925 degrees Cent.) for 4 Hours and Air-Cooled. (Right hand column)  $\times 100$ .

All of the Specimens, Normal as well as Abnormal, Show Coalescence of the Cementite. This is shown in all of the Following Figures of the Specimens Heated at Successively Lower Temperatures. It Should be Noted that there Appears to be Less Free Cementite in the Normal Specimens than in the Abnormal Specimens.

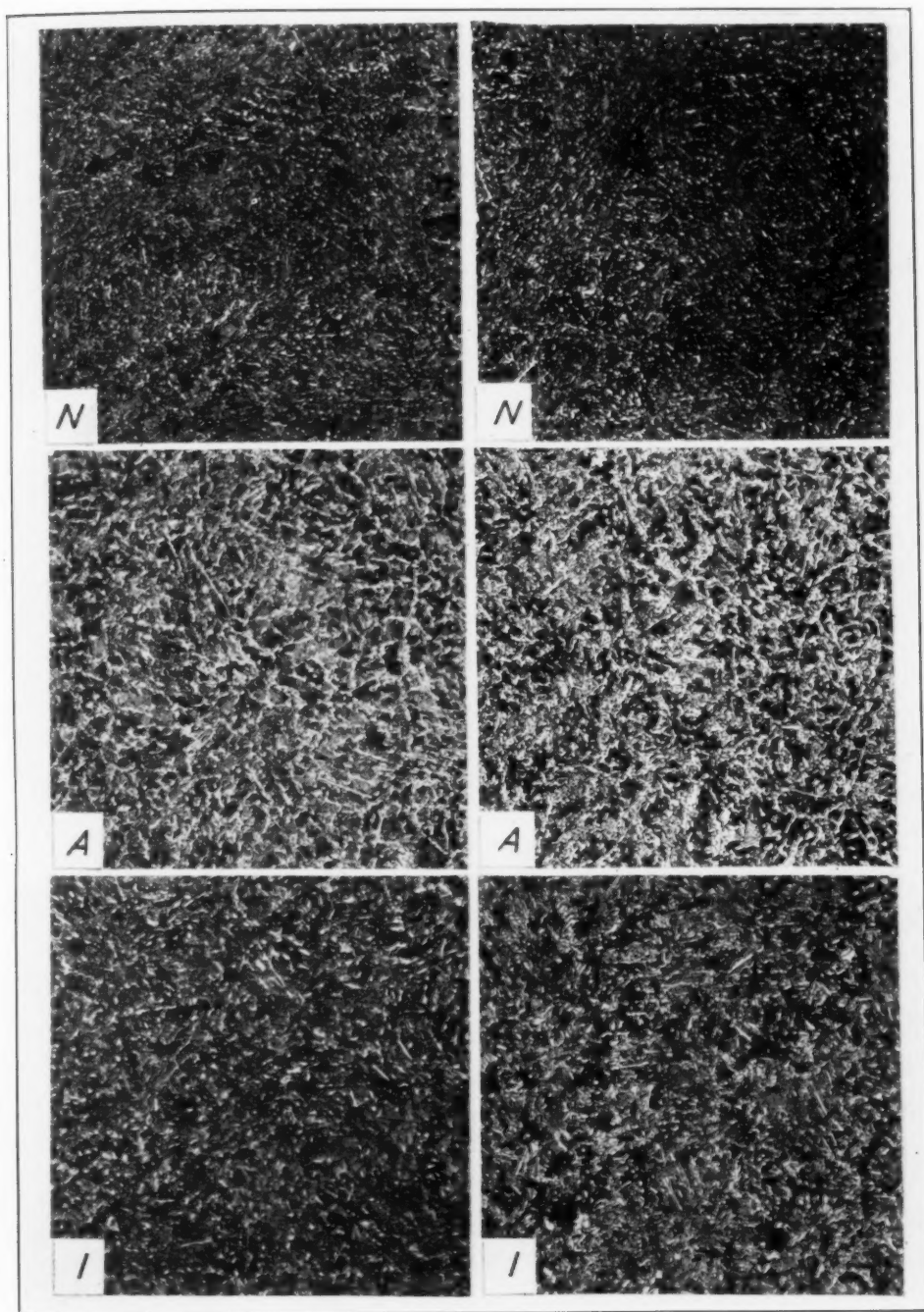


Fig. 8—Photomicrographs of Specimens Heated at 1650 degrees Fahr. (900 degrees Cent.) for 4 Hours and Air-Cooled. (Left hand column)  $\times 100$ . Fig. 9—Heated at 1610 degrees Fahr. (875 degrees Cent.) for 6 Hours and Air-Cooled. (Right hand column)  $\times 100$ .



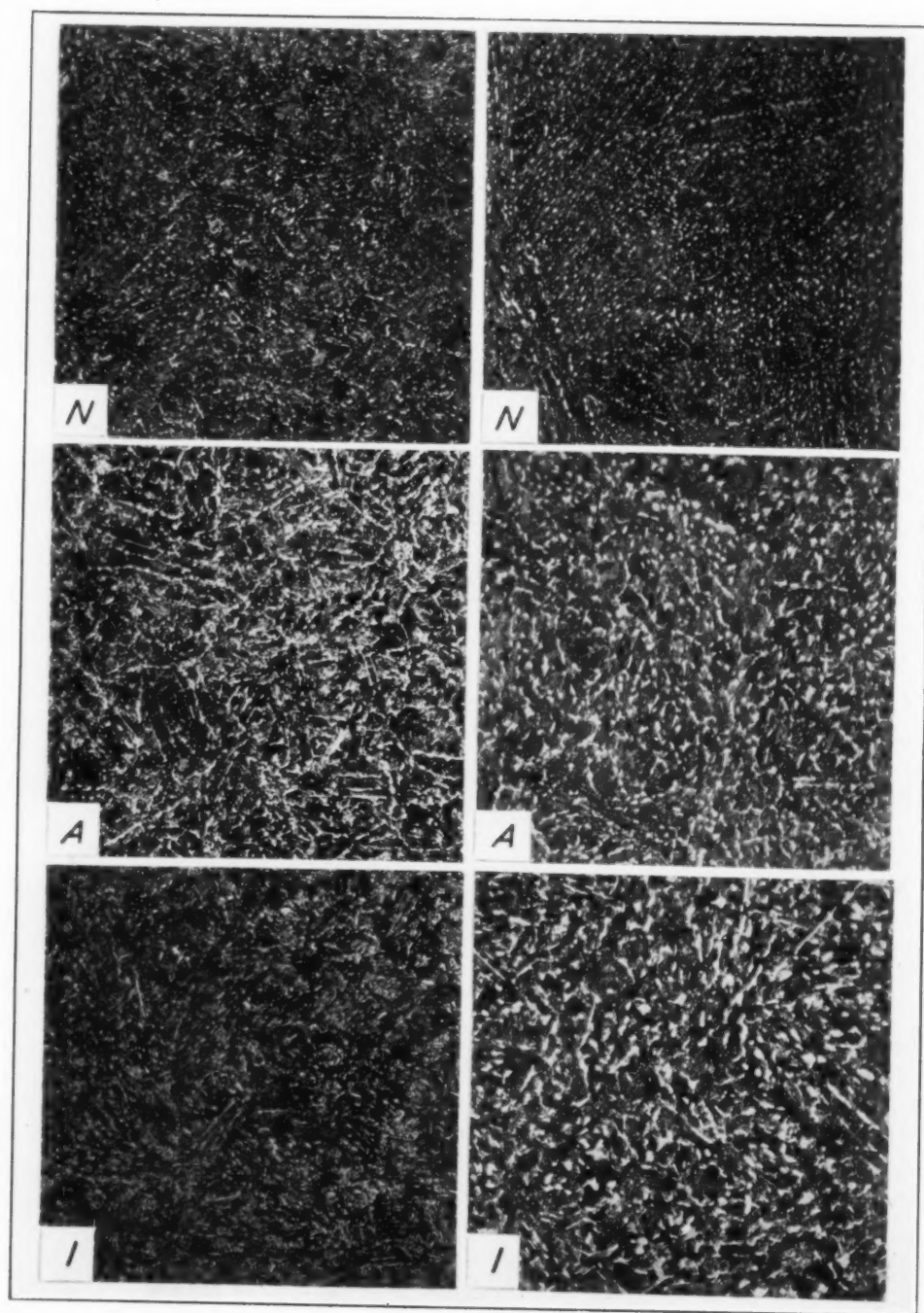


Fig. 10—Photomicrographs of Specimens Heated at 1560 degrees Fahr. (850 degrees Cent.) for 6 Hours and Air-Cooled. (Left hand column)  $\times 100$ . Fig. 11—Heated at 1520 degrees Fahr. (825 degrees Cent.) for 6 Hours and Air-Cooled. (Right hand column)  $\times 100$ .

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tures after heating at lower temperatures down to 1520 degrees Fahr. (825 degrees Cent.) all show coalesced cementite in the normal as well as the abnormal specimens.

From the above it is hard to escape the conclusion that upon heating to a temperature at which all of the cementite is dissolved no coalesced cementite is formed in normal or abnormal steel, but that at a lower temperature at which all of the cementite is not dissolved coalesced cementite is formed in both types. As shown in Fig. 6 upon heating to 1740 degrees Fahr. (950 degrees Cent.) coalescence of the cementite appeared in the abnormal specimens but not in the normal specimens. The fact, therefore, that in a certain temperature range coalesced cementite is formed in abnormal steel but not in normal steel would indicate, that cementite is less soluble in the austenite of abnormal steel than in the austenite of normal steel. We are not concerned here with the cause of abnormality which is still in doubt. As may be noted in Figs. 7 to 11 the abnormal specimens seemed to contain more free cementite than the normal specimens. This also would appear to indicate that cementite is less soluble in abnormal steel than in normal steel. The factor of the concentration of the cementite by diffusion in the austenite as mentioned above may also play a part, however, in the formation of coalesced cementite, and as shown in Fig. 5 such concentration of cementite appears to take place more readily in abnormal steel than in normal steel.

It is evident that considerable further work would be needed to fully clarify this question and perhaps with the renewed interest in the subject, that will doubtless be aroused by this paper such work will be undertaken.

**Written Discussion:** By H. W. Gillett, Chief of the Metallurgical Division, Bureau of Standards, Washington, D. C.

Not the least important point in this interesting paper by Mr. Grossmann, is the presentation of two alternative iron-carbon diagrams. The incompleteness of the traditional iron-carbon diagram, in which no solubility of carbon in ferrite is shown, has long been of interest to the Bureau of Standards. In 1922 Scott<sup>3</sup> published a series of thermal curves showing that  $A_1$  was detectable at 0.041 per cent carbon, but not at 0.030 per cent carbon and stated that no pearlite islands were detectable in specimens containing 0.030 per cent carbon or less. The solubility of carbon in ferrite at the  $A_1$  temperature was thus tentatively fixed at 0.035 per cent carbon. A more complete series of thermal curves from Scott's early work at the Bureau is shown in Fig. 1.

Yensen<sup>4</sup> pointed out that Pilling had found solubility of carbon in ferrite (absence of boundary cementite) at room temperature up to 0.005 per cent carbon by the use of his nitrobenzol etchant.<sup>5</sup> Yensen puts the room temperature solubility as indicated by various electrical and magnetic

<sup>3</sup>See Grossmann's ref. 18.

<sup>4</sup>T. D. Yensen, "Magnetic and Electrical Properties of the Ternary Alloys of Iron, Silicon and Carbon," *Journal, American Institute Electrical Engineers*, Vol. 43, 1924, p. 562.

<sup>5</sup>N. P. Pilling, "Micrographic Detection of Carbides in Ferrous Alloys," *Transactions, American Institute of Mining and Metallurgical Engineers*, Vol. 79, 1924, p. 254.

tests, at somewhere between 0.006 and 0.008 per cent carbon. This is usually stated as "below 0.01 per cent carbon." Yensen's curve for electrical resistivity shows a break at about 0.02 per cent carbon and for minimum reactivity, breaks at about 0.03 and 0.10 per cent carbon.

Hanemann and Schrader<sup>6</sup> accept Yensen's value of 0.006 per cent carbon for the room temperature solubility of carbon in alpha iron, but state that their hypothetical epsilon iron dissolves up to about 0.10 per cent carbon. Adams<sup>7</sup> says that a second break at about 0.08 per cent carbon in one of Yensen's curves could with equal propriety be taken as the carbon content of ferrite in equilibrium with cementite.

Yamada<sup>8</sup> found only boundary cementite in alloys of electrolytic iron and carbon, at 0.034 per cent carbon, but traces of pearlite at 0.04 per cent carbon, thus checking Scott's figures.

Tamura<sup>9</sup> on the three assumptions, none of which are proved, of zero solubility at room temperature in alpha iron, a solubility of 0.07 per cent carbon at 2707 degrees Fahr. (1486 degrees Cent.) in delta iron and a straight line relation between solubility and temperature, calculates the point P' of Grossmann's Fig. 11 to come at 0.034 per cent carbon.

Sisco<sup>10</sup> says that about 0.05 per cent carbon is soluble in ferrite.

Sauveur and Krivobok<sup>11</sup> found no pearlite by microscopic examination in an electrolytic iron-carbon alloy of 0.06 per cent carbon, hence ferrite, they state, can retain at least 0.06 per cent carbon in solid solution after slow cooling. This last is not in accord with the presence of boundary cementite in slowly cooled alloys of lower carbon content.

Guthrie<sup>12</sup> shows cementite in Armeo iron, said to contain 0.02 per cent carbon and describes one streamer of cementite as "in the act of forming pearlite in a ferrite grain." Lucas,<sup>13</sup> however, also working with Armeo iron said to contain 0.02 per cent carbon, shows cementite without any sign of pearlite formation. The history of these Armeo samples is not given in sufficient detail to allow judging whether the cementite was or was not thrown out at the A<sub>1</sub> point. In a private communication from the manufacturers of Armeo ingot iron, it has been stated that it is their experience that cementite areas as shown by both Lucas and Guthrie are of

<sup>6</sup>H. Hanemann and A. Schrader, "On Martensite," TRANSACTIONS, American Society for Steel Treating, Vol. 9, 1926, p. 169.

<sup>7</sup>E. Q. Adams, Discussion, TRANSACTIONS, American Society for Steel Treating, Vol. 9, 1926, p. 235.

<sup>8</sup>Grossmann's ref. 16.

<sup>9</sup>Grossmann's ref. 19.

<sup>10</sup>F. T. Sisco, "Constitution of Steel and Cast Iron," TRANSACTIONS, American Society for Steel Treating, Vol. 10, 1926, p. 463.

<sup>11</sup>A. Sauveur and V. N. Krivobok, "Dendritic Segregation in Iron-Carbon Alloys," *Journal, Iron and Steel Institute*, Vol. 112, 1925, p. 313.

<sup>12</sup>R. G. Guthrie, "Sample Preparation for High Power Photo-Micrography," TRANSACTIONS, American Society for Steel Treating, Vol. 7, 1925, p. 337.

<sup>13</sup>F. F. Lucas, "Observations on the Microstructure of the Path of Fatigue Failure in a Specimen of Armeo Iron," TRANSACTIONS, American Society for Steel Treating, Vol. 11, 1927, p. 531.

rare occurrence in Armeo ingot iron and seldom if ever occur in anything but very slowly cooled material. Hot-rolled bars and air-cooled sheets almost never show grain boundary cementite, while very slowly cooled box annealed sheets will sometimes show it to a very small extent. The manufacturer's ladle analyses of Armeo ingot iron show this material to be in general of the following typical analysis: carbon 0.013 per cent, manganese 0.017 per cent, phosphorus 0.005 per cent, sulphur 0.024 per cent, silicon, trace.

Whiteley<sup>14</sup> examined the solubility of carbon in ferrite and found that a steel of the Armeo type, with 0.035 per cent carbon, 0.02 per cent manganese, 0.026 per cent sulphur, 0.006 per cent phosphorus on heating to 1330 degrees Fahr. (720 degrees Cent.) and quenching, showed a little undissolved cementite remaining. He states that the maximum solubility (at 720 degrees Cent.) is certainly slightly less than 0.035 per cent carbon, and takes the figure 0.03 per cent. He found that the solubility was small at 1165 degrees Fahr. (630 degrees Cent.) and rose to a maximum at 1330 degrees Fahr. (720 degrees Cent.); that by quenching from above the solubility line the solid solution could be retained and that on tempering, precipitation of cementite took place, with rapid migration of the cementite to the grain boundaries. He states "as the purity of the ferrite is diminished, the solubility is, in all probability, reduced."

Atkins<sup>15</sup> has commented on the bad effect in wire-drawing of boundary cementite resulting from the solution of cementite in the ferrite and its precipitation at the grain boundaries.

Masing and Koch<sup>16</sup> have recently shown clearly that the ability to take cementite into solution in ferrite at 1220 to 1292 degrees Fahr. (660 to 700 degrees Cent.), to retain it in solution by quenching and then to precipitate it on aging or on accelerated aging, results in precipitation hardening just as is the case with duralumin.

By quenching at 700 degrees Cent. and aging at 50 degrees Cent. they were able to increase the hardness of low carbon steels with manganese as shown below:

Sample	A	X
% C .....	0.028	0.044
% Mn .....	0.39	0.44
% P .....	0.025	0.036
% Si .....	0.003	0.005
After quenching from 700 degrees Cent.		
Brinell H. N. ....	102	128
After aging at 50 degrees Cent. for 44 hours		
Brinell H. N. ....	163	185

In a series made up from electrolytic iron, without manganese and with 0.02 to 0.03 per cent phosphorus, 0.007 to 0.008 per cent silicon and under 0.001 per cent sulphur, they found—

<sup>14</sup>Grossmann's ref. 20.

<sup>15</sup>E. A. Atkins, "The Drawing of Steel Wire and its Relation to Qualities of Steel," *Journal, Iron and Steel Institute*, Vol. 115, 1927, p. 443.

<sup>16</sup>G. Masing and L. Koch, *Duraluminartige Vergütung bei Eisen-Kohlenstoff-Legierungen*, *Wiss. Ver. Siemens-Konzern*, Vol. 6, 1927, p. 202.



Sample No. ....	1	5	6	9	7	8
% C .....	0.019	0.033	0.043	0.063	0.087	0.166
After quenching from 700 degrees Cent.						
Brinell H. N. ....	85	79	89	123	127	136
After aging at 50 degrees Cent. for 8 days						
Brinell H. N. ....	92	80	101	207	170	194

A long series of different heating times and quenching temperatures and different aging times and temperatures was studied.

There can be no doubt but that Grossmann's Fig. 11 is the correct type of iron-carbon diagram for the materials studied by Scott, Yamada, Whiteley and Masing and Koch. That is, in all these cases, either the specimens with carbon over 0.04 per cent show pearlite, those below 0.03 per cent carbon do not, or the former show  $A_1$  by thermal analysis, the latter do not; or quenching from below  $A_c$ , followed by aging indicates that there is some solubility of carbon in ferrite at 700 degrees Cent. and much less at room temperature. The writer has been unable to find any thermal analysis curves that show  $A_1$  in the steel with less than 0.03 per cent carbon, which is equivalent to saying that no steel has yet been found to conform to the traditional iron-carbon diagram of Grossmann's Fig. 12.

It may very possibly be that Fig. 12 may hold for an oxygen-free steel of 0.03 per cent carbon or less, and that no such steel has ever been produced. It is well-known that Arceo iron, with carbon from 0.01 to 0.02 per cent shows no  $A_1$ , and equally well known, as pointed out by Herty<sup>17</sup>, that it is high in oxygen, (from 0.03 per cent up to perhaps twice that amount, according to oxygen analyses made at the Bureau of Standards,) but no other material of like carbon content shows  $A_1$  either.

Hatfield<sup>18</sup> has pointed out that no pearlite is present in steel below say 0.04 per cent carbon and that since say 0.05 per cent oxygen is also present, the possibility is raised that the carbon and oxygen may both be present as dissolved CO, instead of as dissolved carbon and ferrous oxide.

If Grossmann's theory is correct, that the presence of dissolved iron oxide moves P, Fig. 12, to P' Fig. 11, then treatment with excess aluminum should eliminate dissolved iron oxide and make a steel follow the diagram of Fig. 12 (unless the presence of alumina has a similar effect on the position of P to that postulated for dissolved iron oxide).

None of the previous workers referred to above, except Whiteley, have commented on the effect of impurities in ferrite. Whiteley concluded that the solubility of carbon in ferrite would be reduced by impurities, while Grossmann concludes that it is increased by dissolved oxygen.

While various workers mentioned above give their carbon analyses to the third decimal place, it is no easy task to positively determine carbon within five in the third decimal, on materials of 0.05 per cent carbon and below, and be sure of the absolute accuracy of the results. If and when the difficulties of determining carbon are overcome, the question of carbon segregation remains. How sure can one be, when a given sample of steel,

<sup>17</sup>C. H. Herty, Jr., Discussion, TRANSACTIONS, American Society for Steel Treating, Vol. 13, 1928, p. 468.

<sup>18</sup>W. H. Hatfield, Discussion—Physical Chemistry of Steel-Making Processes, *Transactions*, Faraday Society, Vol. 21, 1925, p. 272.



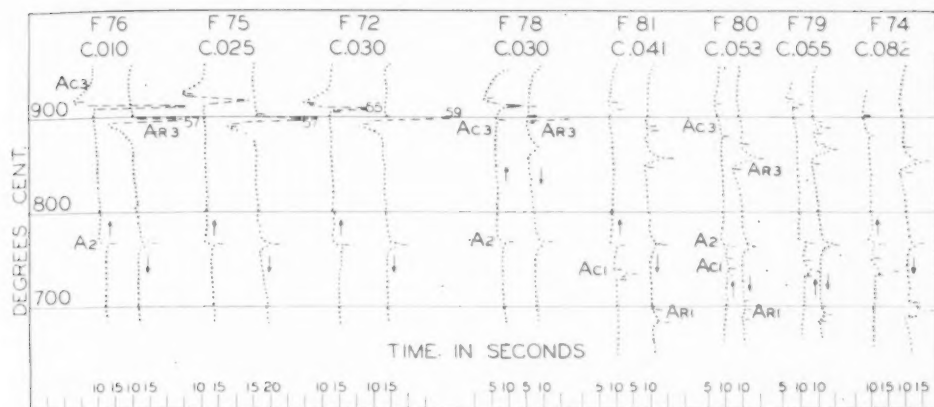


Fig. 1—Inverse Rate Thermal Analysis Curves by Scott. Iron-Carbon Alloys, all Except F-81 made from Electrolytic Iron by Fusion in Argon Vacuum Furnace.

analysis 0.035 per cent carbon, that that is the exact composition of a particular thermal analysis specimen or of a particular field examined under the microscope?

Considerable data on the position of  $P'$ , Grossmann's Fig. 11, have been accumulated at the Bureau of Standards, but the troubles of analysis and segregation have not been sufficiently overcome to allow drawing any final conclusion, and the question raised by Grossmann on the effect of oxygen content, throws the whole matter into still further chaos.

It seems worth while, however, to consider these thermal analysis data if only to bring out the fact that thermal analysis appears to offer a direct method of determining the position of  $P'$  and hence of checking Grossmann's theory.

Inverse rate thermal analysis as used at the Bureau<sup>19, 20</sup> is sufficiently sensitive to show very clearly the weak  $A_2$  point of ferrite and will show a recognizable  $A_1$  point even though the thermal effect is much less than that at  $A_2$ . Of course  $A_1$  might be present but too weak for detection by thermal analysis, nevertheless the carbon content at which  $A_1$  disappears in ferrite of varying oxygen content should give at least a good qualitative conception of the effect of oxygen on the solubility of carbon.

In 1914-1919, H. Scott made thermal analysis of the specimens listed in the following table. The curves for these are shown in Fig. 1.

Spec.	C	Mn	Si	S	P	Cu	$A_1$ Detected
F.74	0.082	.033	.013	.009	..	..	Yes
F.79	0.055	.013	.005	.012	..	Tr.	Yes
F.80	0.053	.033	.007	.019	..	.024	Yes
F.81	0.041(?)	.24	.003	.037	..	..	Yes
F.78	0.030	..	.004	.003	..	..	No
F.72	0.030	..	.005	.006	..	..	No
F.75	0.025	.03	.007	.027	..	.088	No
F.76	0.010(A)	..	..	..	..	..	No

A = electrolytic iron.

<sup>19</sup>Use of a modified Rosenhain furnace for thermal analysis, H. Scott and J. R. Freeman, Jr., Bureau of Standards Scientific Paper No. 348, 1919.

<sup>20</sup>A recording chronograph for the inverse rate method of thermal analysis, H. J. French, Bureau of Standards Technical Paper No. 230, 1923.

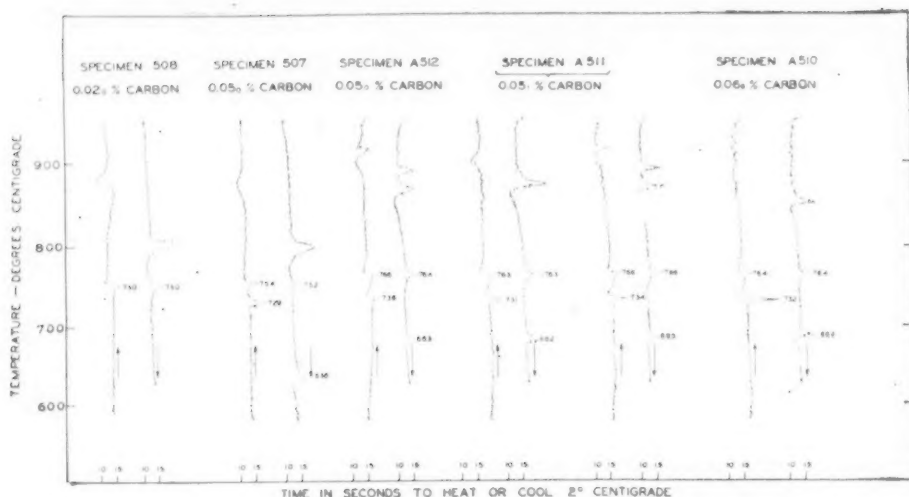


Fig. 2—Inverse Rate Thermal Analysis Curves by Wymore. Iron-Carbon and Iron-Carbon-Manganese Alloys, made from Electrolytic Iron by Fusion in Arsem Vacuum Furnace.

All specimens except F.81 were from 2-lb. heats made from electrolytic iron in an Arsem vacuum furnace, sufficient material being available for analysis so that the carbon figures are probably fairly dependable. F.81 was a small specimen higher in manganese, obtained from Howe and Levy and as insufficient material was available for check analysis, the carbon content is doubtful.

This work indicated that with vacuum-fused electrolytic iron, the solubility of carbon in ferrite at the temperature of the  $A_1$  point lies between 0.03 and 0.05 per cent carbon.

In 1924 Miss I. J. Wymore, in cooperation with J. R. Cain, then research associate for the American Rolling Mill Company, ran thermal analyses on more Arsem furnace melts, including some to which manganese was added in considerable amount.

Number	C	Mn	Si	S	P	Cu	$A_1$ Detected
501	.02 <sub>0</sub>	.35	.01	.01 <sub>4</sub>	.00 <sub>3</sub>	.02 <sub>8</sub>	No
508	.02 <sub>0</sub>	1.23	.12	.02 <sub>1</sub>	.00 <sub>3</sub>	.02 <sub>0</sub>	No (A)
507	.05 <sub>0</sub>	1.06	.05	.01 <sub>2</sub>	.00 <sub>3</sub>	.02 <sub>2</sub>	Yes
A512	.05 <sub>0</sub>	.03 <sub>0</sub>	..	.00 <sub>8</sub>	..	..	Yes
A511*	.05 <sub>1</sub> (B)	.03 <sub>1</sub>	.05	.01 <sub>1</sub>	..	..	Yes (C)
A510**	.06 <sub>6</sub>	.03 <sub>3</sub>	..	.01 <sub>5</sub>	..	..	Yes

A = Microscopic examination showed patches of cementite in some localities, other localities being free from it.

B = Two agreeing analyses by Bureau of Standards; A. R. M. Co. analysis .033.

C = Microscopic examination showed pearlite in appreciable amount, on the whole much less uniformly distributed than is shown in Figure 5.

\*Oxygen by Ledebur method, 0.027% } see B. S. Tech. paper 261,

\*\*Oxygen by Ledebur method, 0.030% } Table I

Various other specimens of Arsem melts, with varying manganese and of commercial ingot iron, all running 0.01 to 0.01<sub>5</sub> per cent carbon were examined, none of them showing  $A_1$ . Nothing was available in this series between 0.02 and 0.05 per cent carbon.

Thermal analysis curves for some of these are shown in Fig. 2. The

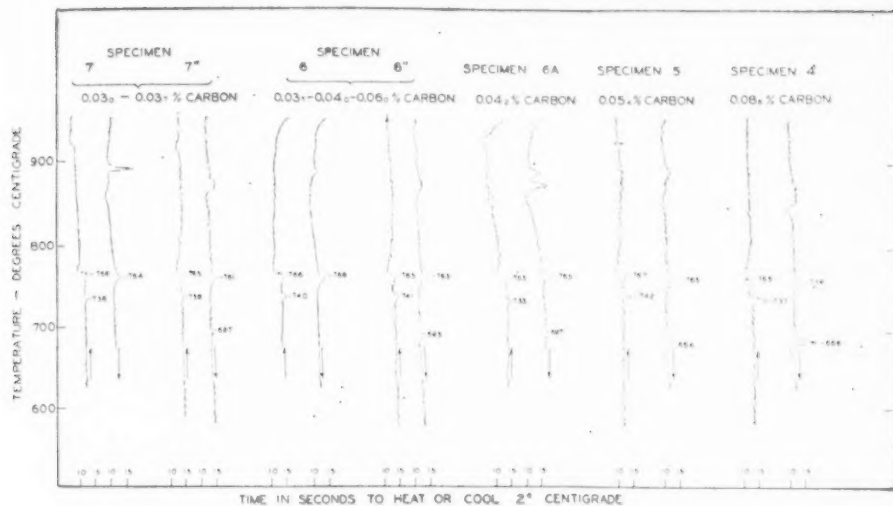


Fig. 3—Inverse Rate Thermal Analysis Curves by Wymore. Ladle Test Ingots Treated with Aluminum.

series served to show that  $A_1$  is clearly present at 0.05 per cent carbon and not detectable below 0.02 per cent carbon irrespective of the degree of deoxidation by manganese. The effect of high manganese on  $A_1$  is clear in 507 and 508.

In the hope of securing specimens in the range 0.02 to 0.05 per cent carbon a series of ladle test ingots taken during the refining of a heat (312179) of ingot iron was studied, the test ingots being killed with large amounts of aluminum (say 2 to 5 grams per lb. of steel).

#### Ladle Tests—Aluminum Treated

No.	C	Mn	Si	S	P	Cu	$A_1$ Detected
4	.08 <sub>8</sub> (A)	.15	..	.03 <sub>4</sub>	.01 <sub>1</sub>	.03 <sub>2</sub>	Yes
5	.05 <sub>4</sub> (A)	.09 <sub>6</sub>	..	.02 <sub>1</sub>	.00 <sub>5</sub>	.03 <sub>4</sub>	Yes
6A	.04 <sub>2</sub>	..	..	..	..	..	Yes
6	.03 <sub>7</sub>	.07 <sub>5</sub>	..	.03 <sub>4</sub>	.01 <sub>2</sub>	.03 <sub>5</sub>	{Yes on heating
7	.03 <sub>0</sub>	.05 <sub>1</sub>	..	.02 <sub>9</sub>	.01 <sub>6</sub>	.03 <sub>1</sub>	{Yes and no on cooling. (D)
8	.03 <sub>7</sub> (C)	.03 <sub>6</sub>	..	.03 <sub>1</sub>	.01 <sub>1</sub>	.03 <sub>5</sub>	{Yes and no on heating. (D)
9	.02 <sub>2</sub> (A)	.03 <sub>6</sub>	..	.03 <sub>1</sub>	.01 <sub>1</sub>	.03 <sub>5</sub>	{Yes and no on cooling. (D)
10	.02 <sub>2</sub> (A)	.02 <sub>7</sub>	..	.02 <sub>3</sub>	.00 <sub>6</sub>	.03 <sub>1</sub>	No

A = A. R. M. Co. analysis of test ingot.

B = B. S. analysis of test ingot 0.06%. A. R. M. Co. analysis of test ingot 0.04%. B. S. analysis of one thermal analysis specimen after run (2 gr. sample) 0.03%.

C = B. S. analysis of thermal analyses specimen after run 0.03%. A. R. M. Co. analyses of test ingot .03%.

D = Several different specimens tested, with variable results.

Fig. 3 shows some of the thermal analysis curves taken on this series.

The behavior of some specimens of 6 and 7 in showing  $A_1$  on some curves and not on others is probably due to slight variation in carbon content. Microscopic examination showed the distribution at room temperature of precipitated cementite in No. 7 to be irregular, cementite being undetectable in some localities and present in segregated "streaks" in others. Fig. 6 shows an average field.

Another ladle test sample 6A from heat 46373, also aluminum treated,

on which Bureau of Standards analysis showed 0.04<sub>2</sub> per cent carbon, gave a thermal analysis curve with a very weak A<sub>1</sub> and only a hint of Ar<sub>1</sub>. It would appear that in these three specimens we must either be just at the point P' of Grossmann's Fig. 11, or else be just on the edge of detectability of A<sub>1</sub>.

Thus in aluminum-treated test ingots A<sub>1</sub> was first detected at 0.03<sub>5</sub> to 0.04<sub>5</sub> per cent carbon.

Other test ingots from Mr. Cain not treated with aluminum and presumably high in both dissolved and precipitated iron oxide were then tested, carbon being determined both by the Bureau of Standards and the American Rolling Mill Company.

A-3		A-4	
.05 <sub>3</sub>	B. S.	.03 <sub>1</sub>	B. S.
.04 <sub>7</sub>	ARMCO	.03 <sub>8</sub>	ARMCO
Av. .05 <sub>0</sub>		.03 <sub>4</sub>	

The A-3 specimen showed A<sub>1</sub> while the A-4 specimen did not.

Another series of test ingots from Mr. Cain not treated with aluminum, was examined by H. C. Cross in 1926. Since carbon segregation was feared, the ingots were forged to 3/4-inch bars and a number of chemical analyses made. The bars were then heated in a vacuum of less than a millimeter of mercury for 7 hours at about 1740 degrees Fahr. (950 degrees Cent.) in the hope of getting better homogeneity. Previous experience with thermal analysis specimens of about 0.05 per cent carbon heated in vacuum showed no diminution in intensity of A<sub>1</sub> after repeated runs, so the loss of carbon is probably not material. The chemical analysis of the thermal analysis specimens after the run shows somewhat lower carbon, but since the samples were small, only 2 grams, the carbon analysis is probably not as trustworthy as analyses on larger samples.

#### Carbon Analyses

B2	C3	C2	B3
.04 <sub>2</sub>	.03 <sub>8</sub>	.04 <sub>1</sub>	.03 <sub>5</sub>
.05 <sub>7</sub>	.04 <sub>5</sub>	.03 <sub>4</sub>	.02 <sub>9</sub>
.04 <sub>4</sub>	.03 <sub>6</sub>	.04 <sub>2</sub>	.02 <sub>5</sub>
.05 <sub>5</sub>	.03 <sub>5</sub> (A)	.02 <sub>9</sub> (A)	.02 <sub>5</sub> (A)
.04 <sub>5</sub>			
.04 <sub>1</sub> (A)			
Av. .04 <sub>7</sub>	.03 <sub>9</sub>	.03 <sub>6</sub>	.02 <sub>9</sub>
Range .04-.05 <sub>5</sub>	.03 <sub>5</sub> -.04 <sub>5</sub>	.03-.04	.02 <sub>5</sub> -.03 <sub>5</sub>

A = Thermal analysis specimen after run.

#### Ladle Tests—Not Treated with Aluminum

	C	C	Mn	Si	S	P	Cu	A <sub>1</sub> Detected
	Av.	After Run						
A2	.09 <sub>1</sub> (B)	.....	.15	.00 <sub>6</sub>	.04 <sub>1</sub>	.01 <sub>0</sub>	.02 <sub>8</sub>	Yes
A3	.05 <sub>0</sub>	.....	.12	.00 <sub>6</sub>	.04 <sub>4</sub>	.00 <sub>9</sub>	.02 <sub>8</sub>	Yes
B2	.04 <sub>7</sub>	(.04 <sub>1</sub> )	.06	.01 <sub>2</sub>	.02 <sub>7</sub>	.00 <sub>4</sub>	..	No
C3	.03 <sub>9</sub>	(.03 <sub>5</sub> )	.11	.00 <sub>3</sub>	.02 <sub>0</sub>	.00 <sub>7</sub>	.05	No
C2	.03 <sub>6</sub>	(.02 <sub>9</sub> )	.08	.01 <sub>2</sub>	.02 <sub>7</sub>	.00 <sub>6</sub>	.05	No
A4	.03 <sub>4</sub>	.....	.10	.00 <sub>4</sub>	.04 <sub>1</sub>	.00 <sub>8</sub>	.02 <sub>4</sub>	No (A)
B3	.02 <sub>9</sub>	(.02 <sub>5</sub> )	..	..	..	..	..	No
A7	.02 <sub>8</sub> (B)	.....	.07	.00 <sub>4</sub>	.03 <sub>1</sub>	.00 <sub>5</sub>	.02 <sub>9</sub>	No
B4	.02 <sub>1</sub> (B)	.....	.03	..	..	..	..	No

A = Microscopic examination of the thermal analysis specimen after the run showed a very few islands of pearlite, very sparsely distributed—see Fig. 7.

B = Armo analysis only.

Fig. 4 shows thermal analysis curves for this series.

Thus in ladle test ingots *not* treated with aluminum,  $A_1$  was *not* detected at 0.03<sub>5</sub>—0.04<sub>5</sub>% C.

In no case cited above, whatever the condition of deoxidation, has  $A_1$  been found with a carbon content below 0.03 per cent and it is present in every case when the carefully determined carbon content is 0.05 per cent or above. (In B. S. Tech. paper 230 thermal curves are given for an 0.06 per cent carbon, 0.24 per cent copper steel which show no  $A_1$ , but without special care in getting sufficiently accurate carbon determinations a carbon

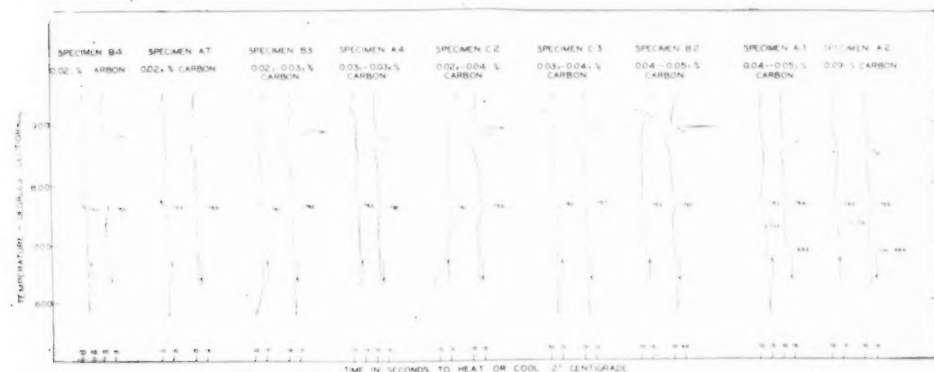


Fig. 4—Inverse Rate Thermal Analysis Curves by Cross. Ladle Test Ingots Not Treated with Aluminum.

figure reported as 0.06 may easily be really below 0.05 so this should not, without further evidence, be taken to indicate that copper affects the solubility of carbon in ferrite. This is mentioned because Grossmann's specimen A contains 0.25 per cent copper). While it may reasonably be argued that there is a limit of sensitivity in the thermal analysis method so that  $A_1$  may occur without being detected, yet the relative location of the point P' in Grossmann's Fig. 11 should be discernable by this means. However, Fig. 7 shows the thermal analysis specimen  $A_1$  of 0.03<sub>5</sub>–0.03<sub>5</sub> per cent carbon which contains an island of pearlite shown by the arrow. Thermal analysis was not sensitive enough to pick up  $A_1$  in this. The thermal analysis specimen of B2, indicated by analysis to contain 0.04 per cent carbon, is free from pearlite as shown in Fig. 8.

Since the carbon content at which one might hope to show the effect of oxygen upon solubility of carbon in ferrite seems to lie in the narrow range 0.03<sub>5</sub>–0.04<sub>5</sub> per cent carbon, it is obvious that to satisfactorily apply thermal analysis to the problem requires (a) a material of extreme uniformity in both carbon and oxygen content and (b) carbon and oxygen analyses accurate to one unit in the third decimal place.

If the carbon analyses for the specimens used in the thermal analyses given above can be accepted, the conclusion seems indicated that in ladle test ingots taken before the steel is finished and presumably containing considerable dissolved iron oxide, no  $A_1$  is detectable in 0.04 per cent carbon



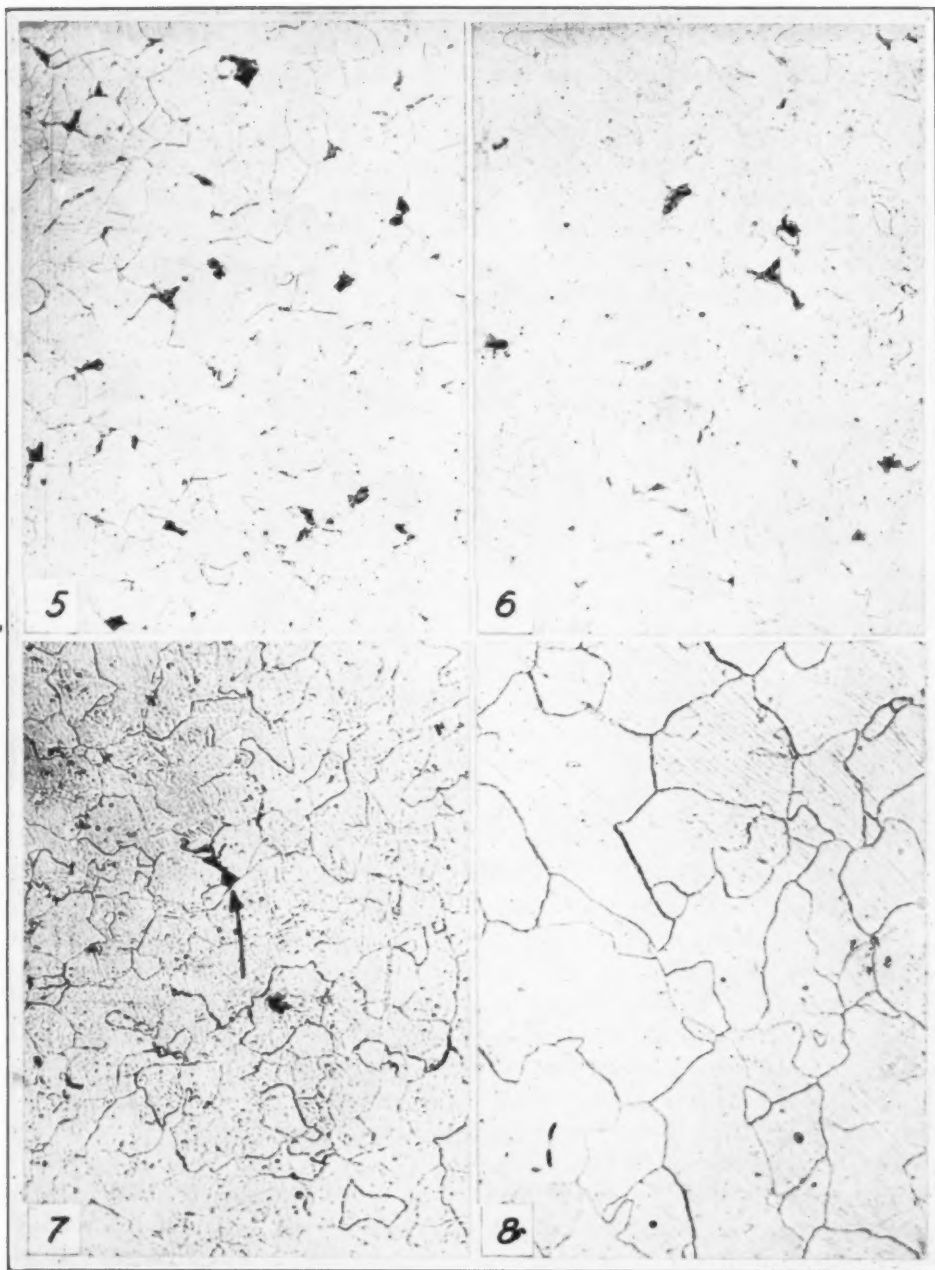


Fig. 5—Average Microstructure of Thermal Analysis Test Specimen A511 after the Run. The Material was not so uniform as the Micrograph might Indicate. This contains about 0.05 per cent carbon. Note the Intensity of A1 in Fig. 2. Fig. 6—Average microstructure of Thermal Analysis Test Specimen of Aluminum Treated Ladle Test Ingot No. 7 After the Run. Other Localities Showed Much More Carbide in Streaks, While Still Others Were Free from Carbide. Analysis Indicates This Material to Contain About 0.03 to 0.035 per cent Carbon. Fig. 7—Average Microstructure of Thermal Analysis Specimen of Ladle Test Ingot A4, not treated with Aluminum. This Material is more Uniform than that of Fig. 6. Analysis Indicates it to Contain 0.03-0.035 per cent Carbon. Fig. 8—Average Microstructure of Thermal Analysis Specimen from Test Ingot B2—not Treated with Aluminum. No Pearlite was Found. Analysis Indicates this to Contain 0.04 per cent Carbon. All Specimens Etched in Picric Acid,  $\times 100$ .

steel, while in more thoroughly deoxidized material (F.81 with 0.24 per cent manganese, 6, 7 and 6A with much aluminum added),  $A_1$  is detectable. This would mean that in the presence of dissolved iron oxide  $P'$  is nearer the right of the diagram and in its absence, nearer the left. That is, it would appear that carbon may be more soluble in the presence of iron oxide as Grossmann's theory calls for.

It would be necessary, before this conclusion could be definitely drawn, to have available material of known and uniform carbon content and known and uniform oxygen content, in the carbon range 0.035-0.045 per cent and make further tests upon it.

These requirements are not sufficiently well met by material so far available and we must be content for the present with pointing out a method of testing Grossmann's hypothesis.

One more bit of evidence from thermal analysis may be presented. Epstein and Rawdon<sup>21</sup> show that in steel of 0.11-0.14 per cent carbon both killed and effervescing, the regular ingot poured after the heat was finished, and with no mold additions, was normal, while ingots treated with aluminum or ferrovanadium in the mold were abnormal. Thermal analyses, (the curves for which are given in the paper), showed  $A_1$  to occur at a very slightly higher temperature, and to give sharper, more intense arrests, occurring more nearly at one temperature, with a longer time of arrest, in the abnormal steels. This suggests that cementite forms more readily at  $A_1$  in abnormal than in normal steel. This would appear to bespeak, if any difference in solubility exists at all, a lower solubility for carbon in the ferrite of abnormal steel, and to support Grossmann's theory since his results indicate that soluble oxygen (iron oxide) favors solubility of carbon in ferrite, hence if oxygen is abstracted from dissolved iron oxide and brought into combination with aluminum as  $Al_2O_3$ , (presumably insoluble in ferrite or at least less soluble than iron oxide), the aluminum-treated ferrite should have a lower solubility for carbon.

But it seems likely, when carbon in excess of the solubility limit as  $A_1$  is present, and  $A_1$  therefore appears, that inclusions, such as  $Al_2O_3$  or other microscopic or submicroscopic oxides or other compounds present in abnormal steel, might serve as nuclei for precipitation of cementite and the formation of pearlite, so the behavior of the 0.11-0.14 per cent carbon steel might be ascribed to this "trigger action" as fairly as to changes in dissolved oxygen. This seems plausible since the difference in thermal analysis curves of normal and abnormal steel appears on cooling, but not on heating.

About all one can conclude is that the location of  $P'$  is a most interesting problem about which we know all too little.

**Written Discussion:** By Dr. O. E. Harder, Professor of Metallography, University of Minnesota, Minneapolis, Minn.

I welcome the results of Mr. Grossmann's researches. I consider that he has been most fortunate in introducing highly accurate oxygen deter-

<sup>21</sup>S. Epstein and H. S. Rawdon, "Influence of the Character of the Steel on the Results Obtained in Casehardening—Normal and Abnormal Steel"—Forthcoming Bureau of Standards Technologic Paper.

minations in connection with his investigation. I had considered this determination of importance in connection with our own work with normal and abnormal steels, but did not have available apparatus for making determinations such as Mr. Grossmann has made. His work will help materially in clearing up some of our troublesome problems under normal and abnormal steels.

I note, however, that in some cases he refers to the diffusion of cementite in austenite or in ferrite and I question the advisability of using this expression. I think there is quite satisfactory information available to indicate that carbon atoms are within the gamma iron lattice and it seems unreasonable to me to think of such a complex particle as is usually pictured for cementite moving through the lattice. It is much simpler to think of the carbon atom simply shifting its attachment by any bonded valence which it may have with adjacent iron atoms. The same discussion probably applies to the diffusion of oxygen. It probably does not move as a group represented by such a formula as  $\text{Fe}(\text{O})_n$  but the oxygen simply shifts the bonding valences whatever they may be to other iron atoms and by this mechanism moves through the space lattice.

Mr. Grossmann has not reported any determinations of the relative rates of diffusion of carbon and oxygen in the steel during the carburizing reaction. Our researches indicate that carbon diffuses in steel more rapidly than oxygen. In one of our papers, (TRANSACTIONS, American Society for Steel Treating, Vol. XIII, p. 961, Fig. 17) we showed a slide which appeared to represent the diffusion of carbon through an oxidized outer zone in sufficient amount to produce a normal structure beyond the oxidized zone. Mr. Grossmann could probably determine this relation by carburizing pieces of steel, removing definite amounts of surface and then making oxygen determinations on the material removed.

I agree with Mr. Grossmann's theory that ferrite is a solvent for carbon because I can conceive no other mechanism whereby a troostitic structure, or a sorbitic structure produced by reheating a quenched steel, can be made into so-called "granular pearlite" or spheroidized cementite. This must involve solution of the carbon from the cementite, presumably the smaller particles, and the movement of the carbon through the ferrite and its final deposition on the large cementite particles. This phenomenon is a slow reaction, as all will recall who have experimentally produced the larger particles of spheroidized cementite. This phenomenon affords a method or means of determining the relative rate of transfer of cementite through the ferrite. For example, if Mr. Grossmann should take quenched steels, some of low oxygen content and others of high oxygen content, and then reheat them to temperatures below the critical and note the relative rate of agglomeration, he might get some valuable information bearing on his problem.

In a paper, (TRANSACTIONS, American Society for Steel Treating, Vol. XIII, p. 961) presented last year we suggested a somewhat different mechanism for the formation of the massive ferrite adjacent to the cementite, and I am still inclined to consider that mechanism quite probable and to

my mind more feasible than the one which has been suggested by Mr. Grossmann. The paper which we have presented this year on the solubility of carbon in normal and abnormal steels seems to support our original mechanism; that is, oxygen in the austenite shifts the Acm line to the left particularly at lower temperatures. This shifting of the line causes cementite to form at higher temperatures for a given carbon content than in the case of normal steels, and once this has started it depletes the adjacent austenite of its carbon content. The adjacent austenite is probably depleted to such an extent that the Acm line intersects the A<sub>3</sub> line to the left of the eutectoid point. The next reaction then is the formation of ferrite, massive ferrite, in those areas of austenite which were most depleted. This reaction, however, then continues until the remaining austenite has been brought up to something approaching the eutectoid concentration at which temperature pearlite forms. The diffusion of carbon through austenite is undoubtedly a reaction that takes place with greater rapidity than the diffusion through ferrite. This is one of the principal reasons for questioning the mechanism proposed by Mr. Grossmann.

**Written Discussion:** By Albert Sauveur, Gordon McKay Professor of Metallurgy and Metallography, Harvard University, Cambridge, Mass.

Mr. Grossmann's important paper is illuminating and will be appreciated. In view of the evidence he offers, it can hardly be denied that in the ordinary pack carburizing of low-carbon steel, oxygen as well as carbon is absorbed by the steel. If I follow the author correctly, even in the presence of a considerable amount of dissolved oxygen, pearlite is formed in the hypereutectoid layer but the ferrite constituent of the pearlite contains both oxygen and some cementite in solution. On further cooling (now in the alpha range) the cementite plates of the pearlite diffuse out of the pearlite and join the hypereutectoid cementite, in some instances resulting in the complete transformation of what was once pearlite into ferrite. This mechanism is attributed to the presence of oxygen in the ferrite of the pearlite. It is not clear to me why oxygen and carbon dissolved in the ferrite lamellae should cause the cementite lamellae to leave the pearlite and be merged with the hypereutectoid cementite. I should appreciate it very much if the author would explain this at greater length. I should also like to ask him to tell us, if he can, the percentages of oxygen in his electrolytic iron both before and after case hardening.

The phenomenon under consideration is a sort of divorcing between the ferrite and cementite of pearlite and recalls spheroidizing which also occurs in the alpha range and makes us wonder whether a steel rich in dissolved oxygen would spheroidize more readily than one poorer in that element. It also brings to mind the fact that it is very difficult to introduce carbon in iron or steel in vacuum, the presence of oxygen being apparently necessary. It is generally believed that the oxygen present in the carburizing box produces some carbon monoxide (CO), that the latter breaks up into C and carbon dioxide ( $2\text{CO} = \text{C} + \text{CO}_2$ ), that the freshly formed carbon does the carburizing work ( $\text{C} + 3\text{Fe} = \text{Fe}_3\text{C}$ ) and that the resulting carbon dioxide is transformed again into carbon monoxide ( $\text{CO}_2 + \text{C} =$



2 CO), thus completing the cycle. The possible part played by oxygen in carburizing, to which Mr. Grossmann calls our attention, may lead to a modification of these views.

**Written Discussion:** By John D. Gat, metallurgist of American Sheet and Tin Plate Company, Pittsburgh.

Mr. Grossmann's paper brings to us the latest developments of metallurgical thought in regard to a most interesting set of phenomena united under the general term "abnormality of steel." He has separated the work of many investigators from the chaff of preconceived ideas and insufficiently well proven facts and succeeded in the presentation of the substance of many a bulky investigation in a few well chosen words. To any one interested in this peculiarity of steel the paper of Mr. Grossmann should be the first, and at the same time the best stepping stone for further research.

Unfortunately, however, in presentation of the results of his own investigations he was not liberal enough in illustrating conclusions reached with more details enabling the reader to follow his reasoning and to agree with his deductions by easier stages than it can be done by reading the paper alone. One would like to ask questions, and many of them, which, with his kind permission I shall try to do.

One would like to know, for example, which other considerations besides the results of oxygen analysis by vacuum fusion method lead him to the assumption of the solubility of the gas in steel under conditions of industrial carburization? From Table I it can be seen that the specimens selected belonged to wide open steels with high oxygen content and undoubtedly possessed all the characteristic features of very abnormal steel. From the same table it follows that the increase in oxygen content was quite pronounced, in case of specimen B averaging 0.04 per cent, an amount as I recall about ten times higher than can be found in properly deoxidized and well made steels, being entirely sufficient for turning any steel in which it might be present into a very abnormal product. While it is quite easy to follow gradual changes from normal to abnormal structure with almost quantitative accuracy, close estimation of the intensification of abnormal features induced in very abnormal steel by increased oxygen content is associated with great difficulty. Any conclusions reached from this observation and not supported by some additional data possess a strong element of uncertainty.

The point of utmost interest both for the maker and user of steel is the possibility of driving these four hundredths of one per cent oxygen into normal steel by ordinary carburization. Were the specimens selected normal, an analysis before and after carburization and changes in structure recorded photographically will answer the question so that no other comments will be necessary.

An objection can be raised, that as long as carburization is inevitable for developing the features by which one can judge the degree of abnormality, it is impossible to divorce the quality of steel from the effects of carburizing operation. Under ordinary conditions it might be so, but approach-



ing the question from the viewpoint of the whole contemporary metallurgical world one can easily digress from the trodden path of similar investigations and to carburize a specimen under conditions precluding oxygen absorption. A piece of properly degasified carbon, bort, carbonado or even a small bit of diamond, placed on steel and heated to a proper temperature in the same high frequency induction furnace in which these experiments were conducted will produce a perfect specimen ranging in carbon content from cast iron to the original structure of steel. Its hyper-eutectoid zone compared with the structure of the same zone of the same steel but carburized under normal conditions will give the first indisputable indications in the right direction.

One can summarize the evidence presented by stating that vacuum fusion method shows increased oxygen content after carburization and that the amount does not depend on carburizing practice, being a function of the properties of the metal. One is left to surmise that the gas entered in a chemical combination with iron.

The work of previous investigators was connected with solid to solid or solid to liquid contact between the reacting phases where there was a strong possibility of direct alloying. Their conclusions can be extrapolated into the realm of equilibria between a solid and gas only with a considerable caution. A definite proof of actual absorption of oxygen by steel is necessary, otherwise, not wishing to depend on circumstantial evidence, no matter how convincing it might appear, one may be inclined to think that the gas is not absorbed, but merely adsorbed.

The results of work of Dr. Yensen and his collaborators definitely proved that in low carbon steels, with which we are dealing in this case, almost half of the carbon content can be removed by heating specimens in vacuum at about 600 degrees Cent. which strongly suggests that this element is present as an adsorbed gas. Oxygen adsorption is well known, it is sufficient to recall its increasing content in the same sample accompanying the reduction of the size of drillings used in determination. Can carbon monoxide of the carburizing process act in the same way or does it follow the path of hydrogen in pickling, travelling along grain boundaries and being retained by internal cavities and nonmetallic inclusions has to be demonstrated, though there seems to be some analogy. Both are strongly ionized, both penetrate in the metal reasonably fast while the smaller size of hydrogen atom probably could be counterbalanced by the increase in space lattice caused by elevated temperature of the process. Carbon monoxide is decomposed with the formation of cementite, but how much of it remains unaffected and how the carbon dioxide formed acts is not known, though it strongly bears on the results of analytical determination of oxygen.

Turning now to the evidence furnished to support the theory that the substance at the grain boundaries is pure cementite one can say that it is as good as the present conception of the subject permits it, but it is not conclusive.

Omitting the well known physical properties of the substance we are obliged to draw all conclusions from the action of etching reagents. The

list of them selected by Mr. Grossmann is fairly comprehensive, but our misfortune is that at the present there is not a single reagent which can actually etch carbides. All proposed solutions hardly do more than slightly oxidize their surfaces without giving any information about their crystalline structure or composition. At the present we are groping just outside the knowledge of the structure of carbides being obliged to be satisfied with such crude distinction that a substance etches as iron carbide, chromium carbide or any other carbide. We cannot tell, for example, whether the mesh surrounding pearlitic grains in hypereutectoid steel is composed of pure iron carbide, or half of it is replaced with chromium carbide, manganese carbide or anything else, and besides we do not know which changes in properties are produced by this substitution.

On one hand we are met with a statement, based largely on circumstantial evidence, that the substance in question is pure iron carbide, on the other there is much evidence, though also circumstantial, that it possesses properties quite different from those which are given in the text books and with which we meet daily in practice. At the point of precipitation of cementite it is not handicapped by anything and has both time and space to arrange its molecules in the pattern inherent to the material. In both oxygenated and normal steels the conditions of precipitation are identical with the exception of somewhat higher temperature in the former case. Still the substance crystallizes quite differently. Thin even continuous lines of true cementite are replaced in oxygenated steels by thick bulging curved inclusions often failing to surround completely the crystals from which they precipitated. In the hypoeutectoid zone this cementite unexpectedly exhibits a strong tendency towards crystallization, which normal cementite lacks always arranging itself along grain boundaries or cleavage planes of the crystals. In Fig. 1 the substance refuses to follow the grain boundaries and shows a decided proclivity toward having a definite shape. The shape of the inclusion of Fig. 2 supports this contention. Its unusual insolubility is only too well known. There is, unfortunately, no more evidence bearing on the distinction between true cementite and the substance under investigation, other than on staining. Specimens of normal and oxygenated steel heat tinted until their matrix was a uniform deep blue had their intercrystalline inclusions colored salmon pink in case of oxygenated metal and brick red in deoxidized metal. Etching with a strong solution of sodium hydroxide developed on these inclusions two shades of reddish brown and almost black. (Figs. 3 and 4.) It might be some staining defect, but it is interesting to note that the black fringe was observed on almost all inclusions free from pearlite and never on those in contact with pearlite.

Mr. Grossmann presented the transformations taking place in hypereutectoid zone with his usual clearness, but he did not follow cementite into the hypoeutectoid zone where a closer study of its properties can aid very considerably in the correct understanding of the influence of oxygen.

The only photograph used for illustrating the behavior of cementite in low carbon areas appears to be chosen somewhat unfortunately because

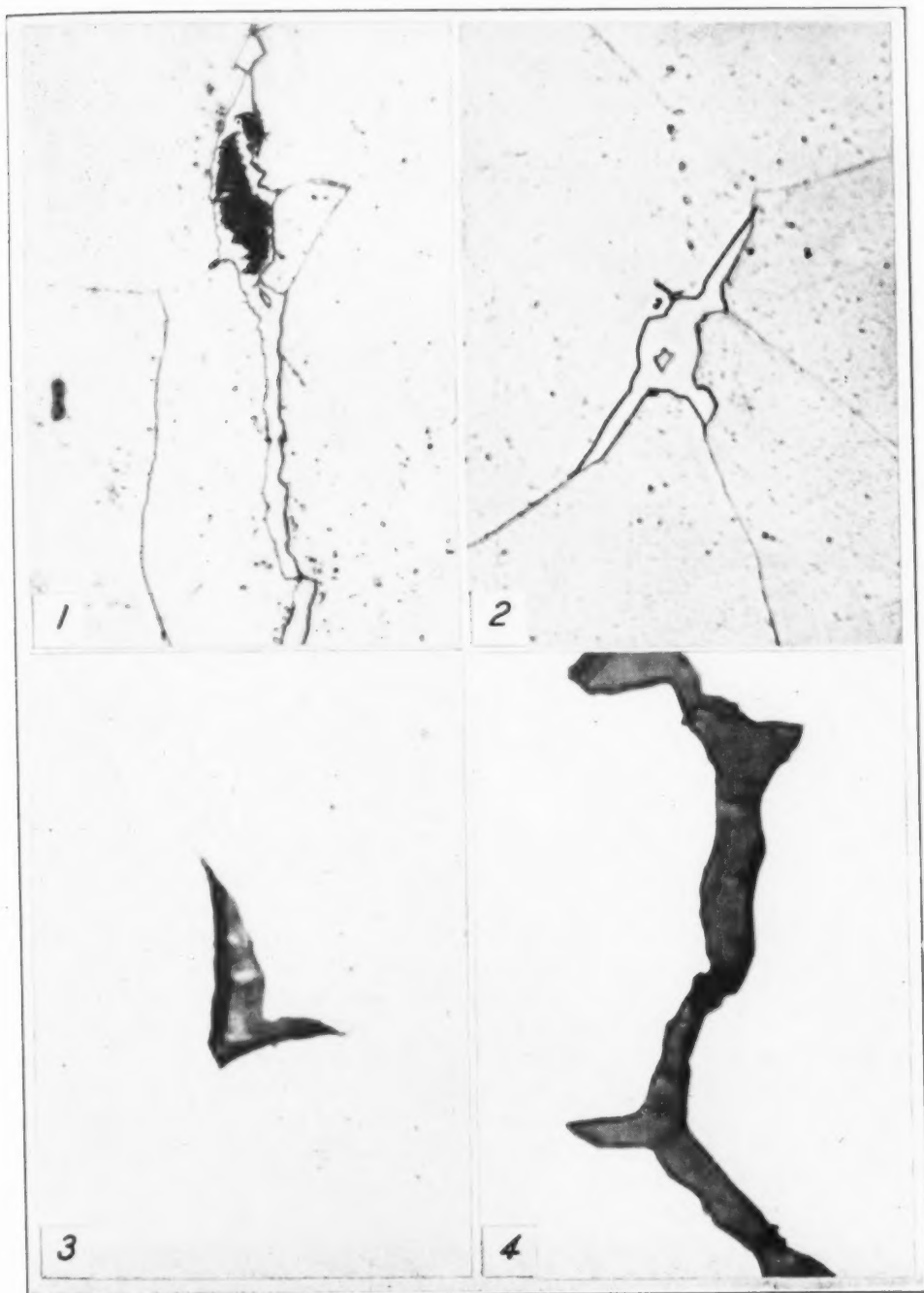


Fig. 1—Inclusion of Cementite in Hypoeutectoid Zone. Note the Rectangular Shape of the Inclusion on the Right. Nital,  $\times 1200$ . Fig. 2—An Inclusion in Cementite in Hypoeutectoid Zone. Nital,  $\times 1200$ . Fig. 3—Cementite in Hypoeutectoid Zone. Boiling Strong Solution of Sodium Hydroxide.  $\times 2000$ . Fig. 4—Cementite in Hypoeutectoid Zone. Boiling Strong Solution of Sodium Hydroxide.  $\times 2000$ .

it represents an exception rather than a rule. Cementite in thin lines following grain boundaries for an appreciable length can be seen in exceptional cases, but usually one sees much bulkier agglomerations of the substance, such as was shown in Fig. 2, Fig. 5 and sometimes approaching the shape of Fig. 6. They, generally, occupy only a small part of the periphery of any given crystal and are often entirely absent at the boundaries. The most common location is the junction of several crystals.

An impression is created that either the surface tension of the substance is so high that it can gather the films rejected from the grains into bodies of much smaller surface, or the precipitation takes place under conditions where we can consider the system as consisting of two insoluble alloys. Surface tension hypothesis can be supported by a rather peculiarly looking Fig. 7 where the substance probably had a chance to form a continuous film but splits itself into a series of fragments. It gains from some work published a year or so ago by Mr. Guthrie on cementite in rimmed steels where the grains of ferrite were outlined by a dotted line formed by this cementite. Mr. Epstein reports that spheroidization takes place much easier in oxygenated steels. The viewpoint of insoluble compounds cannot be accepted in the light of our present knowledge of steel.

In low carbon steels cementite is present alone, but with raising carbon content pearlite begins to appear. The minimum necessary for the appearance of the first traces of it is not accurately determined but is generally accepted as being in the neighborhood of 0.06 per cent largely as the result of work of Yamada. It should be noted that while cementite can be found all over the specimen, provided it is at the grain boundaries, pearlite cannot exist alone being always surrounded with cementite, as is shown on Figs. 8 and 9. The most interesting point is that cementite joins pearlite directly, without any intermediate layer, even in specimens with a hypereutectoid zone of which shows wide bands of ferrite interposed between the grains and the mesh surrounding them. Pearlite, however, does not always crystallize as perfectly as in these photographs, sometimes having certain thickening of cementitic lamellae as is shown on Fig. 10.

The lack of conclusive evidence does not permit us to draw any definite conclusions at the present, but it strongly seems that something is wrong with cementite of oxygenated steels. It does not look like true cementite, and it does not behave like it. Mr. Grossmann's statement that it etches as cementite and therefore is pure iron carbide is the only objection to the assumption that it might be a compound of iron carbide with some compound, presumably oxygen-bearing, which radically changes its properties and relations to iron-carbon diagram.

But is the evidence advanced in support of the purity of cementite convincing? Can we undertake any important work based, for example, on a statement that a certain substance etches as a solid solution, but no means are available for establishing the nature of the components of it? But this is what we are actually doing here.

Until some irrefutable evidence is brought forth establishing the identity of cementites of normal and oxygenated steels all speculations re-



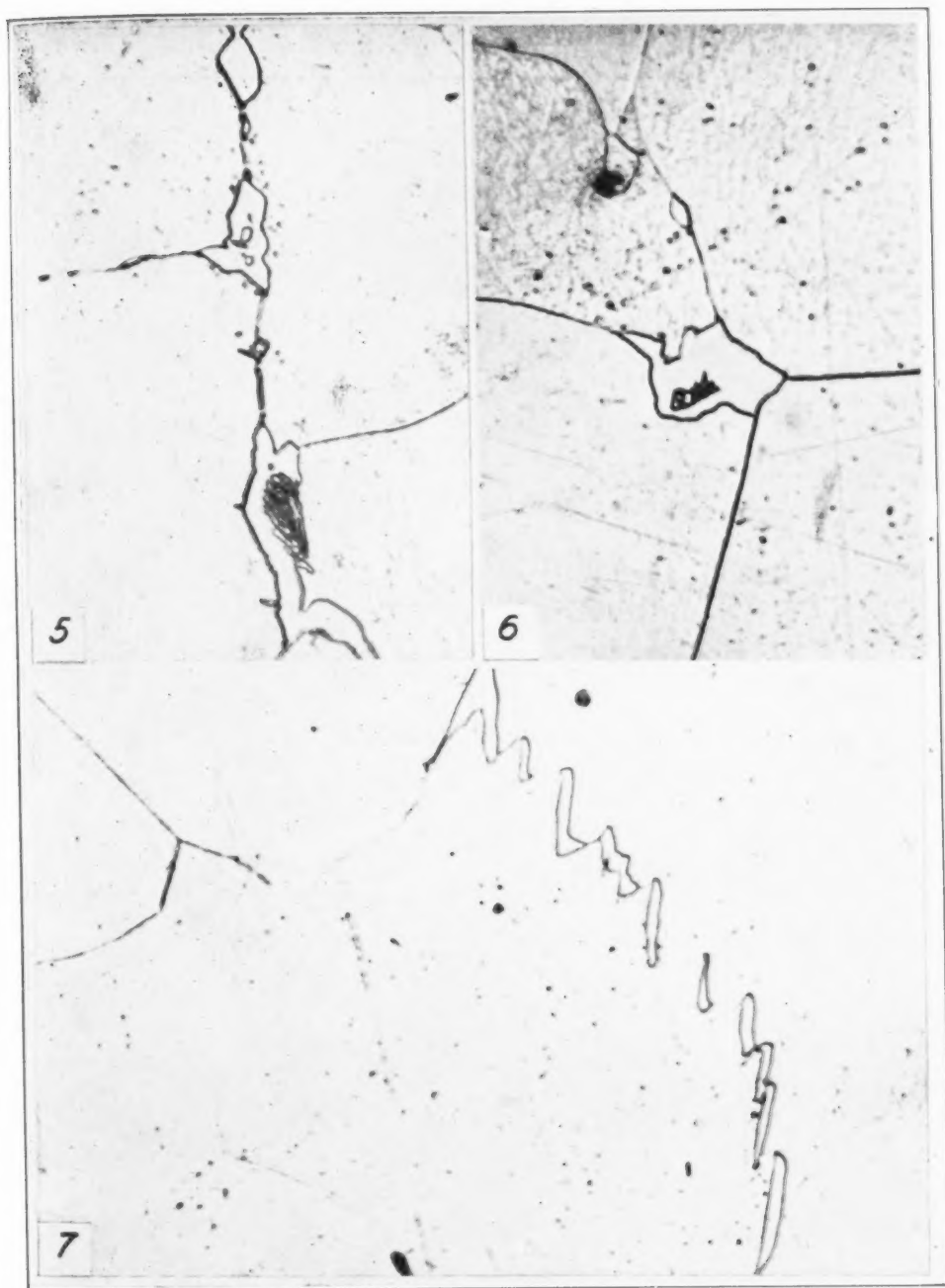


Fig. 5—Cementite in Hypoeutectoid Zone. Nital,  $\times 1200$ . Fig. 6—Cementite in Hypoeutectoid Zone. Nital,  $\times 1200$ . Fig. 7—Cementite in Hypoeutectoid Zone. Nital,  $\times 1200$ .

garding the changes in the iron-carbon diagram will remain under a strong shadow of doubt. We have to assume here that cementite is a true ce-

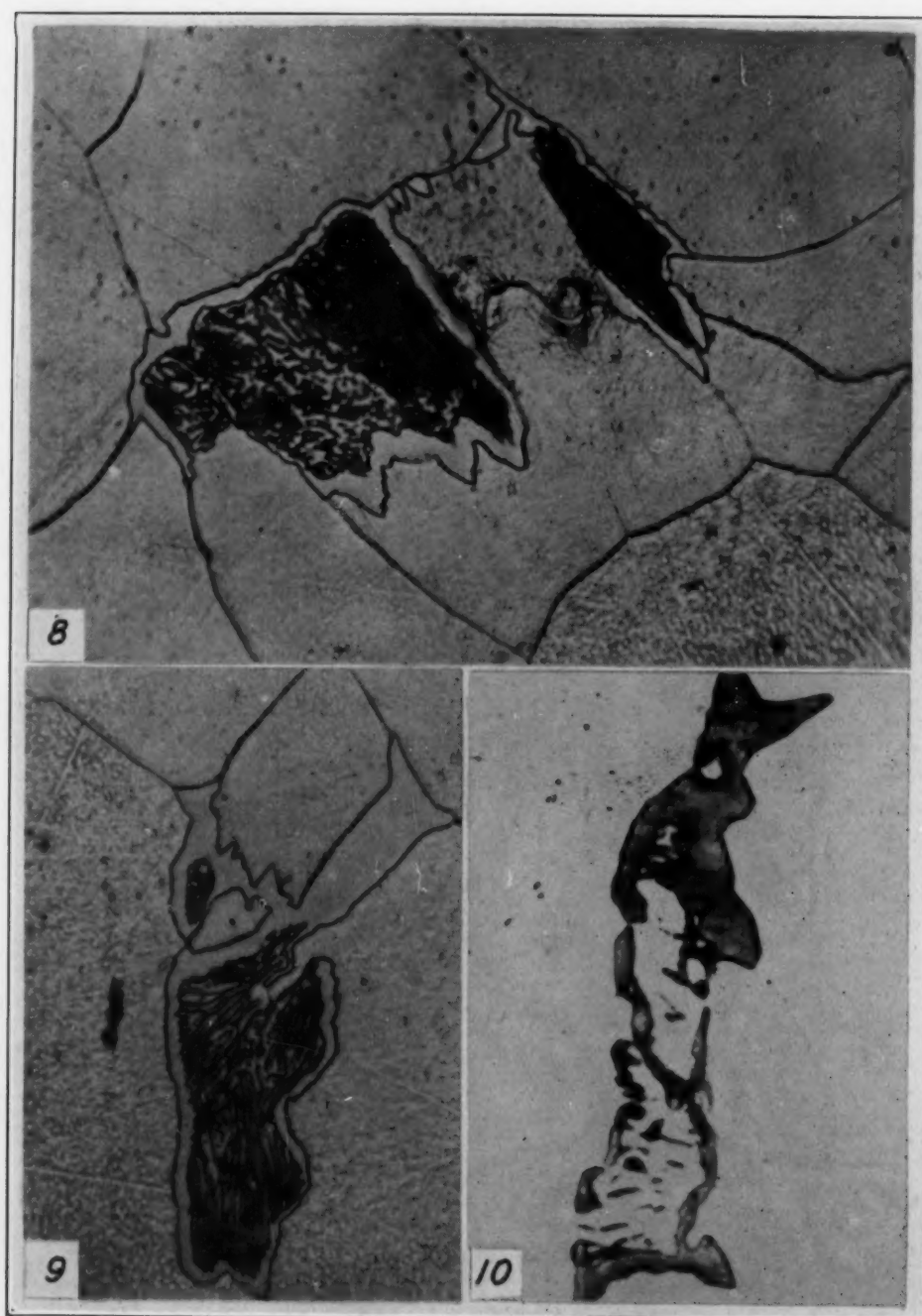


Fig. 8—Pearlite and Cementite in Hypoeutectoid Zone. Note the Width of Cementite Boundaries.  $\times 1200$ . Fig. 9—Cementite and Pearlite in Hypoeutectoid Zone. Note Direct Contact Between Cementite and Pearlite. Nital,  $\times 1200$ . Fig. 10—Pearlite and Cementite in Hypoeutectoid Zone. Boiling Strong Solution of Sodium Hydroxide,  $\times 2000$ .

mentite and all oxygen is concentrated outside of it. With this in mind we may now review the theory proposed.

Three outstanding points which the theory has to explain in this case are the simultaneous existence of ferrite and cementite in the hypereutectoid zone, the presence of cementite at the boundaries of the ferrite of the hypoeutectoid zone and direct contact between pearlite and cementite in the hypoeutectoid zone. The major part of the work described in the paper discussed was conducted on the behavior of steel of hypereutectoid composition, and in the light of the facts furnished by observations of it the acceptance of the diffusion theory is fully justified.

The theory, however, does not give an adequate explanation of the migration of cementitic molecules to the grain boundaries of ferrite. There must be some reason for their travel all across grains to their boundaries. Were the tendency to separate into two phases alone responsible, one could expect here a structure of spheroidized cementite type characteristic of the situation. If ferrite is capable of freeing itself from the last traces of cementite in the hypoeutectoid zone one has to expect in any specimen of oxygenated steel a hypereutectoid zone if not exactly the same as reported by Professor Portevin and Mr. Grossmann in the sample of electrolytic iron, but at least having ferritic bands equal in thickness to one half of the dimensions of ferritic grains, which by no means is the case. The direct contact between pearlite and cementite in hypoeutectoid zone fails to have any reasons for existence on the basis of the theory proposed.

As long as we are dealing with theories would it not be possible, omitting for the present tempting speculations about the nature of the substance hiding under the name of cementite of oxygenated steel, to advance a hypothesis which appears to take care reasonably well of all observations presently recorded besides furnishing a plausible explanation of the progenitor of all work on abnormality, formation of soft spots on quenching.

Instead of increasing the solubility of carbon in iron, oxygen dissolved in it greatly reduces carbon-holding capacity of gamma iron and nullifies it in alpha iron. On reaching the point of precipitation of cementite, strongly shifted to the left as was reported by Dr. Harder and Mr. Johnson, carbon precipitates leaving behind it what would correspond to austenite of eutectoid composition in normal steels, but which in this case contains less than 0.90 per cent carbon.

Further cooling brings usual readjustments resulting at  $A_1$  in austenite of normal eutectoid composition surrounded by the ejected excess of ferrite.

During this period cementite is rejected to the grain boundaries of ferrite which is incapable of holding any carbon in solution and, as the reaction takes place at a comparatively high temperature, the latter is able to take advantage of its abnormally high surface tension and to rearrange itself accordingly. Remaining carbon collects in austenite, rapidly decreasing in volume while ferrite ejected from it immediately unites with the surrounding crystals. At  $A_1$  the austenitic grain is surrounded by ferrite so

that its transformation will result in pearlite surrounded by ferrite as in normal steels. Precipitating from ferrite, austenite carries with it particles of cementite insoluble in the former and which did not have a chance to be rejected to the boundaries. This washing action is purely mechanical because at any given moment austenite carries the full amount of carbon it can have in solution. At the transformation point, crystallization of pearlite rejects cementite to the boundaries.

Quenching of steel from the temperature suitable for normal steel of a given carbon content brings it in oxygenated metal so close to the point of precipitation of cementite that slight retardation in heat transfer, gas and vapor films, etc., results if not in precipitation of cementite, but at least in very small areas in which the carbon of austenite is concentrated. These areas harden normally, but the very much larger masses of impoverished austenite are actually quenched from below the upper critical with the familiar results.

**Written Discussion.** By C. H. Herty, Jr. and B. M. Larsen, Bureau of Mines, Pittsburgh.

Mr. Grossmann has rendered a distinct service in calling attention to the possibility of oxygen absorption in case-carburizing. We say "possibility" because with the uncertainty involved in our present methods for total oxygen determination, even such brilliant and painstaking work as that reported in this paper probably needs some further extension and checking by other investigators. It does not seem entirely certain that all of the refractory oxides present in steel will always be completely reduced in this method. There is a possibility that the higher initial amount of carbon in the carburized pieces may have enhanced the rate of reduction. Also, in cooling in the carburizing box, certain combinations of temperature and gas composition which give oxidizing reactions with iron might exist and produce an oxygen absorption and scaling effect which would not be a necessary element in the reactions at the carburizing temperature. It might be pertinent to ask for Mr. Grossmann's opinion as to the possible cause of the wide variation in amounts of oxygen absorbed in the various samples.

The presence of oxygen is undoubtedly the principal cause of such highly divorced structures as are shown in Figs. 2 and 3. Partly divorced structures have been observed, however, in carburized samples supposedly free of oxygen. Moreover, grain-boundary cementite such as that in Fig. 4 has often been found in ingots killed with an excess of silicon and aluminum alloys, in which the dissolved oxygen content is extremely low.

Both Whiteley's and Tamara's work indicated a probable solubility for carbon in alpha iron of about 0.030 to 0.035 per cent at the  $A_1$  point. Hanson, in discussion of Whiteley's paper, gives a diagram, almost exactly like that shown in Fig. 11. The samples used in these investigations may of course have contained some dissolved oxygen, and Mr. Grossmann's results indicate the need of further study of the alpha solid solution boundary curve in samples known to be very low in oxygen content. From the work of Whiteley, and from the rates of spheroidization of cementite in low-oxygen steels below the  $A_1$  point, it is very probable that carbon has some



small solubility in this range. This solubility value may of course be increased by the presence of dissolved oxygen. It is also possible that the mechanism of the oxygen effect may be that of increasing the rate of carbon diffusion or of decreasing the stability of smaller cementite aggregates, rather than a change in the carbon solubility.

**Written Discussion:** By F. G. Seifing, metallurgist, Michigan State College, East Lansing, Mich.

The author contends that oxygen dissolved in ferrite increases the solubility of the cementite in the ferrite and, therefore, causes the cementite to migrate through the ferrite and to gather with the boundary cementite. According to the mechanism described by the author all the cementite migration occurs at or below the  $A_1$  line.

Fig. 6 undoubtedly shows that the structure was completely austenitic before quenching from 1420 degrees Fahr. (772 degrees Cent.), but the photomicrograph does not reveal variations in the composition of the austenite. It seems to me that nonuniformity of the austenite is an important factor to consider in explaining the diffusion of the cementite from the pearlite to the hypoeutectoid cementite. It is well known that normal steels usually remain normal and that abnormal steels usually remain abnormal when subjected to the common heat treatments and that the banded structure of cold-rolled steels cannot always be removed even with the use of long time anneals at 1850 degrees Fahr. (1010 degrees Cent.). It is reasonable then to suppose that nonuniformity of composition exists in steels even though the structure is entirely austenitic or martensitic. If in the austenite range nonuniformity is possible then it must be taken into account when studying the causes for heterogeneity at temperatures below the austenite range.

I have on different occasions rendered abnormal steels normal with the use of a heat treatment at 2010 degrees Fahr. (1100 degrees Cent.). The resulting normal steel was identical in every respect with other normal steels. A peculiar case was encountered, however, in Armco ingot iron. Different samples of the ingot iron were heated to 2010 degrees Fahr. (1100 degrees Cent.) for different lengths of time varying from  $\frac{1}{2}$  hour to 6 hours, when one piece was quenched in water, one piece cooled in lime, and one piece cooled in the furnace. All the samples were then carburized at 1700 degrees Fahr. (925 degrees Cent.) for 8 hours. This treatment yielded structures which were all very similar and all very abnormal. Several of these carburized pieces were then treated at 2010 degrees Fahr. (1100 degrees Cent.) for 1 hour and cooled in water or lime and then carburized again at 1700 degrees Fahr. for 8 hours which resulted in unusually normal structures in the case.

Possibly the phenomena of all the above steels and ingot iron can be explained in the light of their oxygen content; but I am inclined to think that the oxygen content is not the only factor affecting the abnormality except in so far as the dissolved oxygen affects the temperature at which complete and uniform solution in the austenite takes place.

In 1926 the suggestion was made<sup>22</sup> "that austenite may not be a complete solid solution. From the evidence concerning normal and abnormal steels<sup>23</sup> it seems to me there is a critical solution temperature for the austenite of each steel which temperature is definitely affected by very small quantities of oxygen, aluminum, vanadium, etc.

**Written Discussion:** By Paul D. Merica, Development and Research Department, International Nickel Company, New York City.

The importance of cementite solubility in alpha iron has begun to be better realized within the last few years. It has been noted by Whiteley (1927) and others that cementite is soluble in alpha iron and that its solubility diminishes at lower temperatures below the value 0.03 per cent carbon determined by Whiteley at 1330 degrees Fahr. (720 degrees Cent.). For it should be emphasized that the appearance of cementite at alpha grain boundaries indicates not only cementite solubility, but also that the latter is decreased with decreased temperatures.

The author's attention was caught by this fact, and by the still more striking one that it does not always seem to be true. Apparently cementite solubility in alpha iron sometimes changes with temperature and sometimes it does not, at least according to the circumstantial evidence of structure. In searching for a reason for this variation he has been led to ascribe it to the presence of dissolved oxygen in some form—when oxygen is present the cementite solubility diminishes with temperature, when absent, it does not change. I believe the possibility worth considering that the dissolved oxygen instead of increasing cementite solubility actually decreases it at the lower temperatures—the cementite being normally approximately equally soluble at all temperatures.

I should like to point out the bearing of this very interesting work of Mr. Grossmann upon the problem of temper brittleness. Andrews, Dickie and others who have given much study to this problem, have concluded that there is a precipitation of carbide at alpha grain boundaries in temper brittle steels, which indeed causes the brittleness. They fail to explain, however, why all steels are not equally temper brittle. Yet no fact is more certain than that temper brittleness is not primarily determined by composition (within the usual scope), since of two steels of identical composition, one may be very temper brittle and the other quite free from it.

On the other hand, the late Professor Oberhoffer advised me privately that he had sound evidence that temper brittle steels contained more oxygen than non-brittle ones, regardless of other elements.

Mr. Grossmann's hypothesis therefore reconciles these two rather isolated lines of thought and investigation. There is a carbide precipitation in temper brittle steels, because they contain dissolved oxygen which causes

<sup>22</sup>Frederick G. Seifing, "Abnormal vs. Normal Tool Steels," Michigan Engineering Experiment Station Bulletin No. 5.

<sup>23</sup>O. E. Harder and W. S. Johnson, "Solubility of Carbon in Normal and Abnormal Steels," TRANSACTIONS, American Society for Steel Treating, Vol. 15, 1928.

Frederick G. Seifing, "Control of Normality in Plain Carbon Steels," Michigan Engineering Experiment Station Bulletin No. 13.

B. M. Larson and A. W. Sikes, "A Note on the Effect of Heat Treatment on Abnormal Case Carburizing Steels," TRANSACTIONS, American Society for Steel Treating, Vol. 14, 1928, p. 355.

the cementite solubility to diminish with temperature. In steels of low oxygen content the cementite cannot precipitate because its solubility remains substantially unchanged.

#### Oral Discussion

O. W. McMULLAN: I was much interested in Mr. Grossmann's paper, and have a question or two that I would like to ask him. He mentions that in steel, showing the split boundary effect, when quenched above the Acm line, martensite was produced without any diffusion of the carbon. I believe it is quite generally conceded that this type of structure is very difficult to harden, it giving soft spots in the carburized piece. Can this carbon diffusion take place so rapidly that it is the failure of the quench to prevent it from taking place that gives these soft spots or must the separation have taken place above the Acm line?

With gas carburized samples, where there is no oxidizing atmosphere, do we not get a similar structure to pack carburizing as regards the split cementite effect?

Another point I might mention is that scaled pieces, when carburized, will show immediately beneath this oxide scale a very normal structure in steels that ordinarily will give a normal structure free from the split cementite effect, the oxygen in scaling either not penetrating the steel or if so having no effect on the normality.

L. J. WEBER: I should like to ask Mr. Grossmann some questions as regards the determination of the oxygen content. In the method that he uses the steel is melted in vacuum, both before and after carburizing, and the oxygen content calculated from the gas given off, which is considered to be carbon monoxide. During the carburizing operation the following factors may effect these results:

1. Absorption of carbon monoxide, nitrogen, carbon dioxide on the surface.
2. The presence of these gases in the voids of the steel.
3. Actual solubility of these gases in the steel.
4. Formation of compounds of iron and carbon monoxide and iron and nitrogen.

Were these factors taken into consideration in reporting the results obtained in these analyses?

In Table I, the difference in the oxygen content before and after carburizing is given at 0.034, 0.042, and 0.028 per cent. On the fifth page it is mentioned that the German investigators reported a difference of 0.078 per cent in the oxygen content of the same material but using samples of different fineness. What does Mr. Grossmann consider to be the accuracy of the results which he obtained in his gas analyses?

#### Author's Reply to Discussion

With regard to Mr. Epstein's and Mr. Jordan's discussion, they have brought some interesting data bearing on oxygen absorption or non-absorption, as indicated in a certain number of tests. The differences between their results and ours can be summarized as follows. They have tested some heats (originally very low in oxygen) and found but a slight change after carburizing, sometimes a slight gain, sometimes a slight loss.

All of our tests made on commercial steels (not reported here) showed substantial gain. One test reported by them, on a material high in oxygen, showed loss after carburizing. We have had one example of such behavior. One test on steel high in oxygen showed no change in their tests. We have generally found a substantial gain. Inasmuch as our analyses showing high oxygen are of the same order of magnitude as those of the Bureau, 0.050 per cent, and since we have checked with the Bureau on our standard sample containing low oxygen, 0.005 per cent, we believe the methods of analysis are in substantial agreement. It seems probable therefore that discrepancies represent different conditions of carburizing, including temperatures. It should perhaps be mentioned that minor differences in method of analysis are still a matter of discussion. One thing seems certain: oxygen does migrate and can be considered soluble at carburizing temperatures. The equilibria should therefore be studied.

I would like to call attention more emphatically, however, to the very significant photomicrographs which they showed regarding the shape of the cementite particles in the hypereutectoid zones of the carburized material. (Epstein and Jordan's Fig. 2.) It seems to me probable that a very considerable portion of the question of normality and abnormality will hinge on the significance of the fact that they show cementite plates rather than rounded particles in the hypereutectoid zone under certain conditions of cooling.

With regard to Dr. Gillett's discussion, that is so complete that there is really nothing further to be said about it. He had already expressed his convictions regarding the solubility point in the iron-carbon diagram which appeared in a recent issue of the *TRANSACTIONS*, American Society for Steel Treating showing solubility of carbon in ferrite placed at about 0.04 per cent. It is interesting that, following the work which he has done and which he has reported on, there have recently been a few further studies in the literature just in the last few months, one of them by Köster which appeared in the discussion of the same paper by Masing and Koch to which Dr. Gillett referred. They studied this same point by making carbon determinations by the Eggerts method on low carbon steels quenched at successively higher temperatures just below the  $A_1$  point. The Eggerts method is presumed to determine only the combined carbon, and when the carbon went into solution upon somewhat higher heatings, there was a decrease in the apparent carbon, indicating the going into solution of some of the carbide.

The paper by Pfeil in a current Iron and Steel Institute preprint refers to an estimation of the amount of dissolved carbon in ferrite by studying the proportion of pearlite in samples of known carbon content. It was found that the pearlite did not represent all of the carbon present.

Thus as further evidence becomes available, it only confirms Dr. Gillett's estimate of the extent of carbon solubility.

There is also a paper by Bramley and Haywood in the current Carnegie Scholarship Memoirs showing the effect of oxygen on aging effects after cold drawing. Since the oxygen affects these properties, it might be as-



sumed to be dissolved oxygen which migrates and so has its effect.

Dr. Harder's point regarding cementite diffusion versus carbon diffusion is of course a controversial one. There is certainly much evidence which points to carbon diffusion as carbon atoms instead of as cementite molecules. I am quite ready to agree with that, although that would not affect the final structures or the conclusions drawn.

In regard to the relative diffusion rate of carbon and oxygen, it seemed probable to us too at first that the carbon would diffuse more rapidly than the oxygen. However, there is some evidence which would indicate that the reverse is true and that the oxygen diffuses ahead of the carbon.

In regard to Dr. Harder's mechanism for the formation of the ferrite in the split cementite structure, we can only point out that this mechanism would suggest that when the cementite has precipitated, the austenite stays depleted in carbon and appears therefore as a ferrite layer. This would suggest that carbon diffusion is slower in high oxygen steels. This is contrary to the evidence offered by Mr. Epstein that diffusion is more rapid in high oxygen steels. Diffusion into austenite seems to be more rapid in the high oxygen steels.

Dr. Sauveur was good enough to bring up several interesting points. With regard to the building up of the large cementite masses, I have for my part subscribed to the view that the cementite particles have an attraction and orienting power for carbon atoms or for cementite molecules in the immediate neighborhood of the cementite particle. This attraction and orienting power would cause a slight depletion of the cementite in the alpha iron in the immediate neighborhood of the cementite particle, resulting in more diffusion of the carbide from the adjacent ferrite layers, the process thus becoming continuous and finally resulting in a complete separation of the phases, given sufficient time.

With regard to the spheroidizing in higher oxygen materials, that suggestion is very interesting and one well worth following.

With regard to Mr. McMullan's question about the martensite structure, it seems probable that the carbon diffusion takes place so rapidly that when the piece is quenched with insufficient speed, troostite forms and soft spots occur. That is, the guilty structure forms in the quench. As regards the gas carburized samples, there may already be so much oxygen in some steels, that the amount is sufficient to cause the phenomena noted. As regards the normal structure just beneath the scale, much further study is needed on the distribution and the amounts of oxygen in different layers.

Mr. Weber mentions various possibilities as to the manner of occurrence of gases in steel. The conclusions offered in the paper are the ones which seemed reasonable on the basis of the data reported, and we can only say that we have no data as yet indicating the occurrence of any compounds of the nature suggested. As regards the variations caused by differences in size of sample, we were very careful to use comparative samples always of the same size, as pointed out in the paper, so that there would be no discrepancy of this sort whatever. The accuracy of the results is considered to be well within 0.002 per cent.



The discussions by Dr. Merica, by Mr. Gat, by Dr. Herty and Mr. Larsen, and by Mr. Sefing are greatly appreciated.

Dr. Merica has made most interesting reference to the subject of temper brittleness, and the field seems indeed to be one for fruitful study. Since temper brittleness is influenced by the rate of cooling from the tempering temperature, the phenomenon would involve the emphasis Dr. Merica has laid on the decrease of cementite solubility with the lowering of the temperature.

In reply to Mr. Gat's discussion, it should be noted that the arguments presented in the original paper are based on two observed facts. (1) The change in oxygen content caused by carburizing indicated that oxygen does diffuse in steel at carburizing temperatures, and is therefore presumably soluble in steel at those temperatures. (2) Steels having a high oxygen content show "abnormal" cementite and pearlite (Figs. 2, 3, 4 and 9) while steels with a low oxygen content show normal cementite and pearlite (Figs. 13, 14 and 15). Only the extreme cases are reported here, but a current paper by K. Inouye (Memoirs of the College of Engineering, Fukuoka, Japan) gives an interesting series of such structures.

It may well be, as argued so ably by Mr. Gat, that the cementite of high oxygen steels is not pure  $\text{Fe}_3\text{C}$ , but contains something else, possibly oxygen. Its behavior in steel reactions is, however, like that of cementite, and it seemed wise to call it cementite, just as we refer to the cementite of low chromium steels, although we know well that there is some chromium present in such cementite.

In considering ferrite, cementite and pearlite, it should be remembered that pearlite is only a transitory structure, and occurs as pearlite only because the separation of cementite and ferrite was succeeded by conditions affording no opportunity for the ferrite and the cementite to agglomerate. (Pearlite is not a phase.) The presence of oxygen seems to give opportunity for agglomeration. It seems difficult to think of austenite rejecting first massive cementite and then massive ferrite above the  $A_1$  point.

Dr. Herty and Mr. Larsen point quite properly to the possibility that the carbon absorbed in carburizing might enhance the rate of reduction during the actual analysis, and so account for the apparent increase in oxygen content. A strong argument against this, however, is the fact that in some steels initially high in oxygen, analysis after carburizing shows a lower value than before, instead of a higher, indicating in this case a loss of oxygen in carburizing. It is not believed that the cooling in the carburizing box acts over a sufficient time to affect the values given. As to the mechanism causing different steels to absorb different amounts of oxygen in carburizing, that must remain a matter for future study.

Mr. Sefing's contention that there may be a lack of homogeneity in austenite seems indeed a reasonable view, and one which must be considered. The anomalous behavior of the Armeo iron could perhaps be explained by a change in oxygen content due to presence of carbon in the second high temperature treatment, though this is only speculation in the absence of data. In any case, much interest would attach to further reports on Mr. Sefing's part.

## SOME CHARACTERISTICS OF PEARLITE IN EUTECTOID RAIL STEELS

By O. V. GREENE

### *Abstract*

*The author gives the results of tests made on rail steels of eutectoid composition. The photomicrographs included show the character of the pearlite in these steels. N. T. Belaiew's method is used for explaining the apparent difference in the nature of the pearlite. The condition of the pearlite is dependent not on the carbon content but on the velocity of cooling through the  $A_{r_1}$ . A definite relation exists between the Brinell hardness and the distance between the lamellae. Equations are developed showing the Brinell hardness as a function of both the distance between the lamellae and the ultimate tensile strength. As the Brinell hardness and ultimate tensile strength increased, the distance between the lamellae decreased.*

FOUR years ago an investigation was instituted to determine if possible the relation of the microstructure to the physical properties of rail steels. The work was continued until March, 1928. The results obtained and the conclusions reached are contained in this paper.

With few exceptions the rail steels bought to our specifications are of about eutectoid composition. Consequently for the purpose of comparison 130-pound R. E. section rails of eutectoid composition only, were investigated. Since we were interested mainly in the rail heads, the photomicrographs, Brinell hardness and physical tests were taken from these portions of the rails. The tests were taken from each rail in exactly the same manner. The location of the tests is shown in Fig. 1. The tensile tests were turned to standard A. S. T. M. 0.505-inch bars. The sections for microscopic examination were removed from the ends of the test bars opposite the ends containing the Brinell impressions, before the bars were tested. The microscopic specimens were cut with

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enough surface and thickness to enable the taking of check Brinell hardness tests after microscopic examination. The chemical analyses were taken from the test bars and the results are given in Table I.

#### THEORY RELATING TO MICROSTRUCTURE

Since the rails were of eutectoid composition and cooled in the air after rolling with no subsequent heat treatment, the struc-

Table I  
Chemical Analysis of Rails

Rail Number	Per Cent				
	C	Mn	P	S	Si
1	0.845	0.78	0.034	0.043	0.23
2	0.835	0.81	0.037	0.035	0.16
3	0.850	0.73	0.020	0.025	0.19
4	0.840	0.71	0.030	0.030	0.22
5	0.850	0.78	0.028	0.044	0.25
6	0.850	0.70	0.025	0.042	0.27
7	0.865	0.73	0.030	0.039	0.25
8	0.855	0.81	0.024	0.045	0.23
9	0.835	0.75	0.026	0.030	0.27
10	0.855	0.73	0.039	0.043	0.23
11	0.845	0.76	0.029	0.045	0.17
12	0.855	0.70	0.023	0.030	0.22
13	0.860	0.74	0.026	0.031	0.25

tures were composed wholly of lamellar pearlite. For the interpretation of the pearlitic sections examined, the author is deeply indebted to Col. N. T. Belaiew's investigations on the inner structures of the pearlitic grain.<sup>1</sup> As his papers and texts are probably not as well known in this country as they should be, a brief review of his method of analysis will be given.

The pearlite grain is considered as an allotriomorphic crystal; that is a crystal the external shape of which bears no relation to its internal symmetry. It is polyhedral in form with its shape depending not on its crystallographic properties but on the conditions of the crystallizing process. The internal structure of the polyhedral grain of pearlite will be composed of alternate layers of cementite and ferrite, arranging themselves parallel to the crystallographic planes of the crystalline mass of the grain. As alpha iron crystallizes in cubes, Belaiew assumes that these layers will arrange themselves parallel to the faces of a cube. During the

<sup>1</sup>N. T. Belaiew "The Inner Structure of the Pearlite Grain". *Journal, Iron and Steel Institute*, 1922, No. 1, Vol. CV, pp. 201-27.

N. T. Belaiew "Crystallization of Metals". University of London Press, Ltd.

process of crystallization the first deposit will form along one of the faces. Later lamellae will orient themselves parallel to the face that has been chosen by the first deposit. If the three-dimensional point of view is kept in mind, it is obvious that a secant plane in a specimen containing such polyhedrons, will cut a great number of grains with the lamellae at various angles. Consequently the coarseness of the pearlite, the size of the cementite lamellae, and the distance between them is dependent upon the position of the secant plane. What is actually seen is the projec-

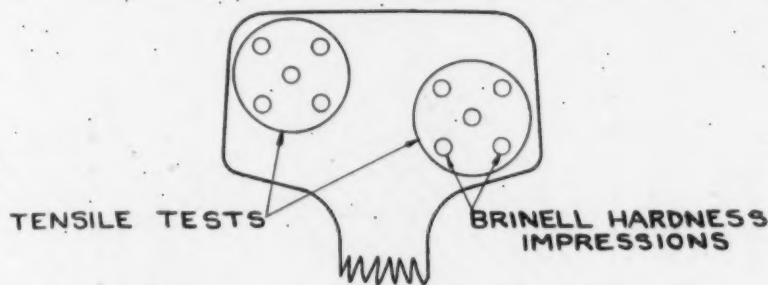


Fig. 1—The Location of the Tests on the Heads of the Rails.

tion of the lamellae on that plane. In order to show the arrangement of the lamellae in space, the stereometry of the pearlite grain has been fully developed by Belaiew.

It is assumed that a sphere is inscribed in a polyhedral grain, and that the lamellae of cementite and ferrite are closely packed parallel with one another and with the first meridian. This is shown in Fig. 2 with the lamellae drawn in only  $\frac{1}{4}$  of the sphere. The lamellae are perpendicular to the equatorial plane and show a series of equidistant parallel lines. If the number of lamellae in the sphere be 200, a quadrant section will contain 100. The quadrant section shown in Fig. 2 shows this on plane ABF with every tenth lamellae drawn. A section ADF is cut at an angle of 60 degrees to the normal plane. Only 50 lines appear on this plane with the apparent spacing of the lamellae considerably increased. Thus the reason is explained for finding on examination that some of the grains have extremely fine laminations while others are very loosely packed giving comparatively coarse ones. Belaiew has developed this more completely as given in Fig. 3. He shows it is evident that a plane inclined to the normal at an angle  $\Phi$  of 26 degrees will show 90 instead of 100 lamellae. At this angle the apparent thickness and distance between lamellae



has decreased by 10 per cent. Sections with the angles of inclination increasing to 37, 45, 54, 60 and 66 degrees show respectively 80, 70, 60, 50, and 40 per cent of the lamellae. At these angles the character of the lamellae appears much the same as it did on the normal plane with the exception that the pearlite is merely less closely packed as the angles increase. Figs. 9 and 13 show

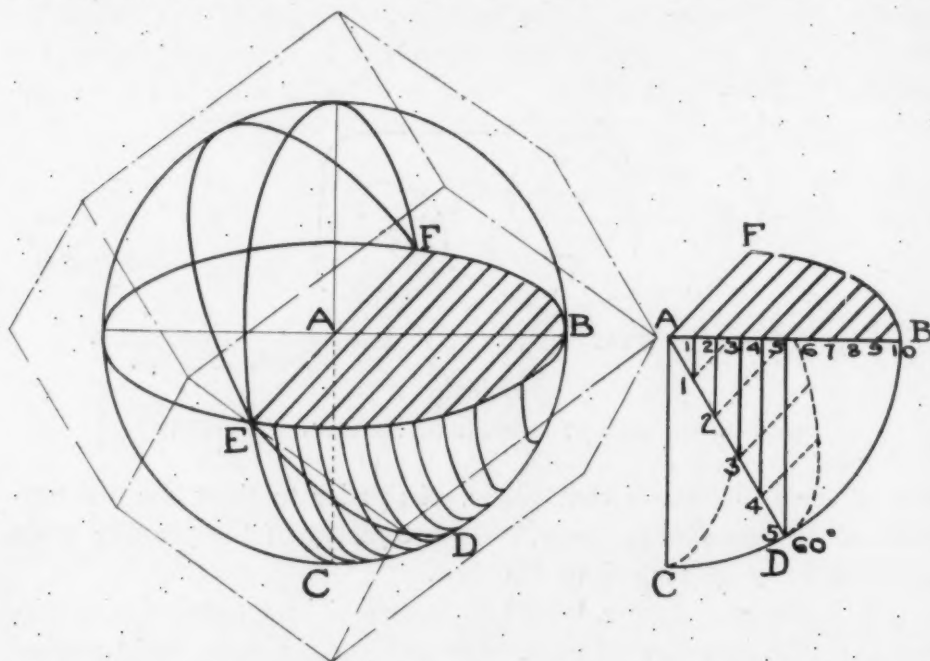


Fig. 2—The Assumed Relations of the Lamellae Contained in a Sphere Inscribed in an Idealized Polyhedral Grain of Pearlite.

the appearance of the lamellae on the normal plane  $\Phi$  equals 0 degrees. The lamellae of Figs. 5, 6, 14 and 16 at various angles up to 67 degrees are of the same nature as those in Figs. 9 and 13 except that they are more loosely packed. Referring again to Fig. 3 it may be noted that on the sections at 73 and 78 degrees only 30 and 20 per cent respectively of the lamellae appear. The pearlite has now become definitely coarse, which may be seen in Figs. 7, 8, 12 and 18. At 84 degrees only 10 per cent and at 87 degrees 5 per cent of the lamellae are visible and their appearance has now become quite peculiar. Note the curved and broken appearance of the pearlite inclined at 84 degrees in Fig. 17. Thus by simple stereometric relations, the difference in the appearance of various grains within the same specimen is accounted for.

Since Belaiew was the first to devise this method of analysis



parts of his discussion and his exact nomenclature with the exception of the angle  $\Phi$ , will be given in the following:

Fig. 4 shows the method he employed for the calculation of the angle of inclination  $\Phi$ . The position of a plane in space is

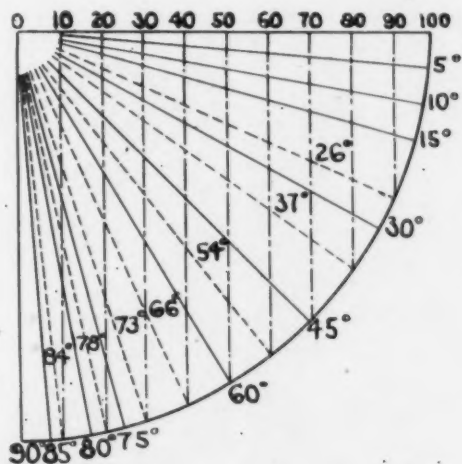


Fig. 3—Quadrant Section of Ideal Pearlite Grain. (From N. T. Belaiew, *Journal of Iron and Steel Institute*, 1922, No. 1, Vol. CV, pp. 201-27).

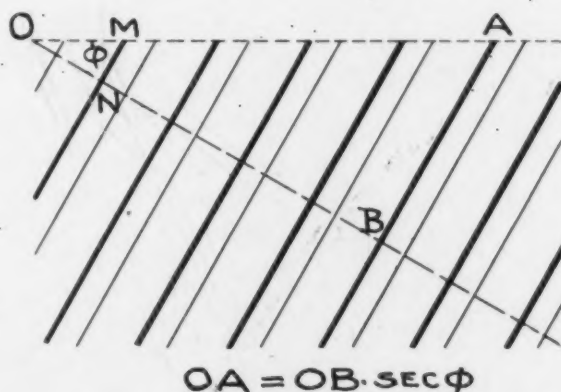


Fig. 4—Transverse Section of Cementite Lamellae. (From N. T. Belaiew, Reference Same as Fig. 3).

determined by two angles. In this case the angle of inclination of the plane to the axial plane, and the angle of deviation between the x or y-axis and the trace of the secant plane on the axial plane. (See Fig. 2). The axial plane coincides with the equatorial plane so the latter is 0, and the angle of inclination is designated by  $\Phi$ .

The distance between two cementite lamellae is designated

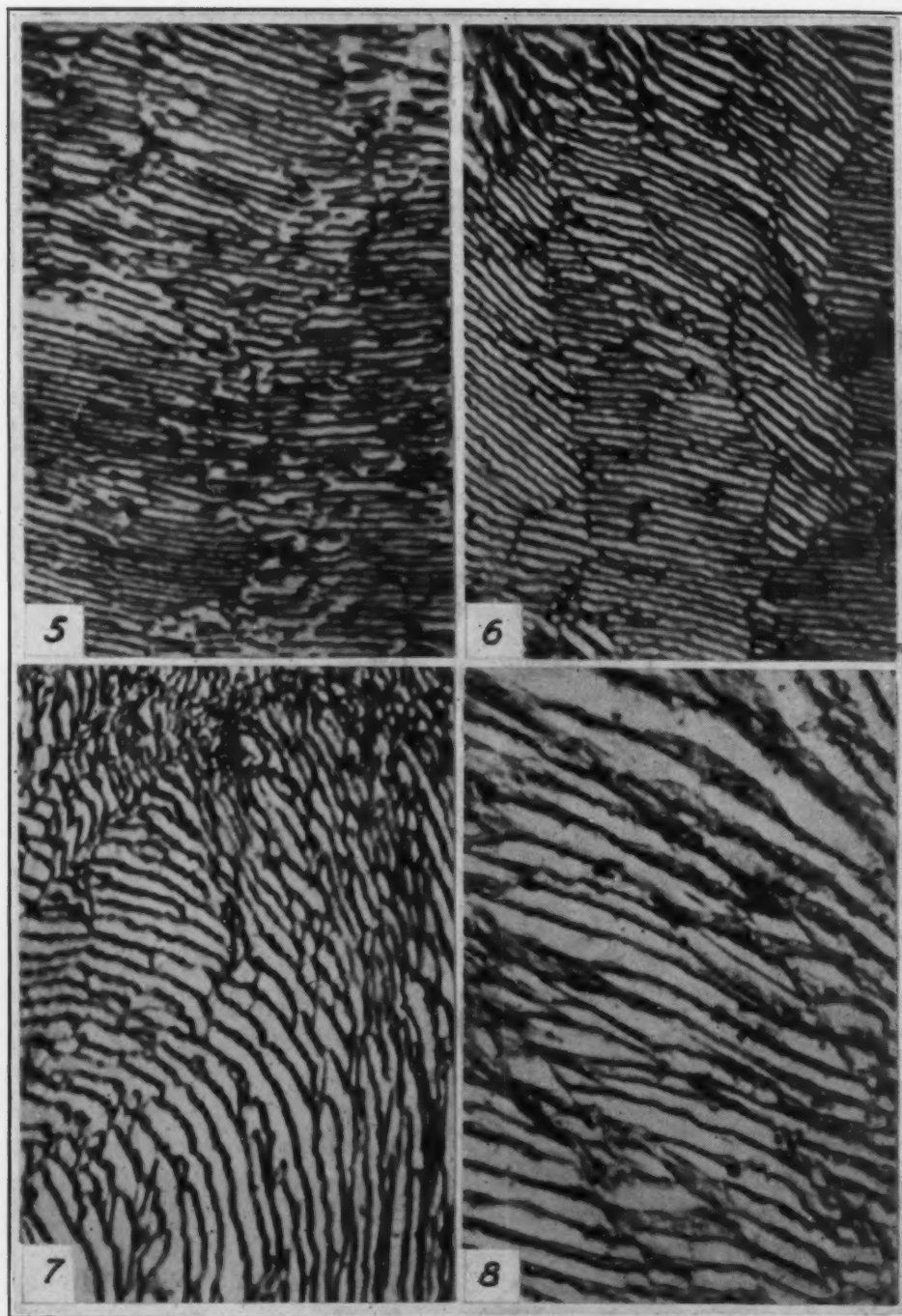


Fig. 5—Photomicrograph of Rail No. 6,  $\Phi = 45$  Degrees.  $\times 2500$ . Fig. 6—Photomicrograph of Rail No. 6,  $\Phi = 61$  Degrees.  $\times 2500$ . Fig. 7—Photomicrograph of Rail No. 6,  $\Phi = 71$  Degrees.  $\times 2500$ . Fig. 8—Photomicrograph of Rail No. 6,  $\Phi = 76$  Degrees.  $\times 2500$ .

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by  $\Delta$ . The apparent distance between them on a section inclined at an angle  $\Phi$  will be  $\Delta_{\Phi}$ . As the distance between two planes is measured on a normal plane and as the angle of inclination of such a plane is 0, the actual distance between two lamellae is denoted by  $\Delta_0$ . In Fig. 4 a series of parallel lamellae are cut by a secant plane at an angle  $\Phi$ . The following are the relations for the triangle AOB.

$$\cos \Phi = \frac{OB}{OA}$$

But  $OB = 5ON$  and  $OA = 5OM$  where  $ON = \Delta_0$  and  $OM = \Delta_{\Phi}$ . Therefore:

$$\cos \Phi = \frac{\Delta_0}{\Delta_{\Phi}}$$

The values for the angle  $\Phi$  of the pearlite in photomicrographs were calculated by the last formula.

In order to determine the  $\Delta_0$ , the specimens were examined at magnifications of 2500 to 5000 diameters which were sufficiently high to resolve the finest lamellae. The whole section of each specimen was carefully examined and the number of lamellae in 2 or 3 centimeters of every fine grain was measured on the ground glass of the microscope. The distance between the lamellae could be then calculated in microns, 1/1000 millimeters ( $\mu$ ). The grains with like values having the smallest distance between the lamellae were taken for the determination of  $\Delta_0$  for the specimen.

$\Delta_{\Phi}$ , the apparent distance between two adjacent lamellae of cementite on a section inclined at an angle  $\Phi$ , can be easily measured and calculated in microns in the same manner.

#### THE PEARLITE OF RAIL STEEL

The photomicrographs shown in Figs. 5 to 19 may be used as an aid in substantiating the foregoing theory. The  $\Delta_0$  for each rail was first determined. The results are given in Table II. The values for  $\Delta_0$  determined as previously described, were quite constant for the whole area of each of the heads of the rails examined. Consequently this may be taken as ample proof that the rails cooled uniformly.

Figs. 5, 6, 7 and 8 of rail No. 6 show the lamellae at angles from 45 to 76 degrees. Figs. 13, 14, 15 and 16 of the various

rails show angles of  $\Phi$  increasing from 25 to 69°. It can be seen that there is very little difference in the character of the lamellae at these angles with the exception that as the angle  $\Phi$  increases, the apparent width of the lamellae increases and the pearlite seems coarser. If Fig. 5 be compared to Fig. 13 it may be seen that for most of the grain in the latter  $\Phi = 49$  degrees and it apparently is the same or somewhat finer than the grain  $\Phi = 45$  degrees in Fig.

Table II The Distance Between Lamellae $\Delta_o$ , for Eutectoid Rail Steel				Table III The Values of $\Delta_o$ , the Distance Be- tween the Lamellae and their Cor- responding Brinell Hardness			
Rail Number	Microns	Centimeters	Lines Per Inch Actual	Rail Number	$\Delta_o$ in Microns	Brinell Hardness	Constant
1	0.338	0.0000338	75000	5	0.266	302	80.3
2	0.309	0.0000309	82400	9	0.270	293	79.1
3	0.292	0.0000292	87000	6	0.276	288	79.5
4	0.276	0.0000276	92000	4	0.276	286	78.9
5	0.266	0.0000266	95500	7	0.278	283	78.7
6	0.276	0.0000276	92000	8	0.280	285	79.8
7	0.278	0.0000278	91400	3	0.292	270	78.8
8	0.280	0.0000280	90600	2	0.309	257	79.4
9	0.270	0.0000270	94100	12	0.334	241	80.5
10	0.340	0.0000340	74700	1	0.338	239	80.7
11	0.338	0.0000338	75000	11	0.338	235	79.3
12	0.334	0.0000334	76000	10	0.340	232	78.9
13	0.376	0.0000376	67500	13	0.376	215	80.8

5. The reason for this is explained by the fact that  $\Delta_o$  for Fig. 13 is  $0.01\mu$  less than that of Fig. 5.

Figs. 11 and 12 show lamellae at about the same angle of inclination. The distance between the lamellae of Fig. 12 at 79 degrees compared to that of Fig. 11 at 80 degrees can be readily understood, for  $\Delta_o$  in Fig. 12 is  $0.048\mu$  greater than that of Fig. 11.

The vast difference in appearance possible between grains in the same specimen is shown by Figs. 9 and 10.

If Fig. 10 be compared with Fig. 11 it may be seen that at angles of  $\Phi = 80$  to 81 degrees the lamellae have assumed a definitely different appearance. The lamellae show a slight tendency to become curved and broken while they are still roughly parallel.

Figs. 17 and 19 show the lamellae with  $\Phi = 84$  and 83 degrees respectively. Here the lamellae are very curved, broken and not definitely parallel as may be expected at these angles of inclination.



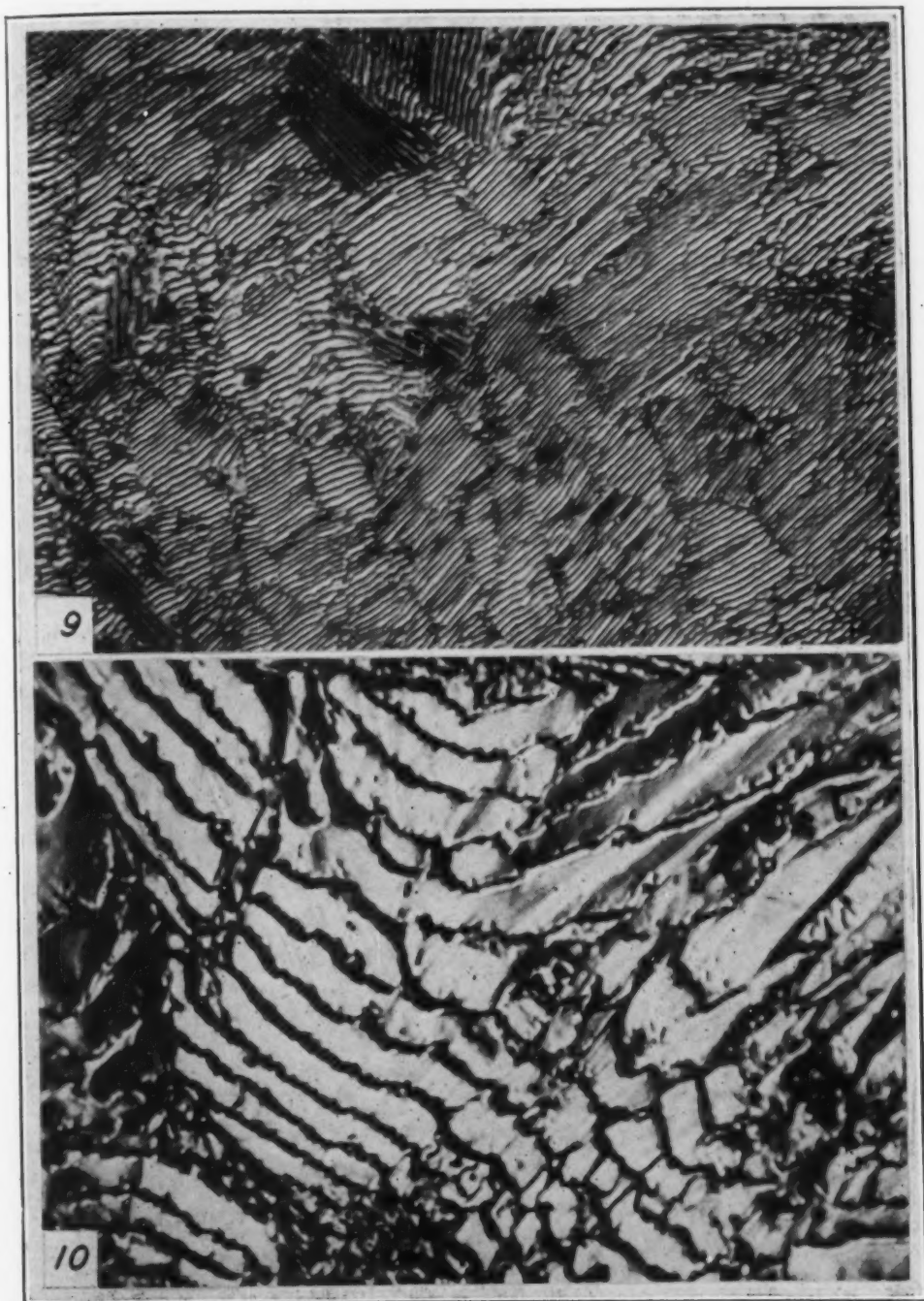


Fig. 9—Photomicrograph of Rail No. 2,  $\Phi = 0$  Degrees,  $\Delta_0 = 0.309\mu$ .  $\times 2500$ . Fig. 10—Photomicrograph of Rail No. 2,  $\Phi = 81$  Degrees.  $\times 2500$ .



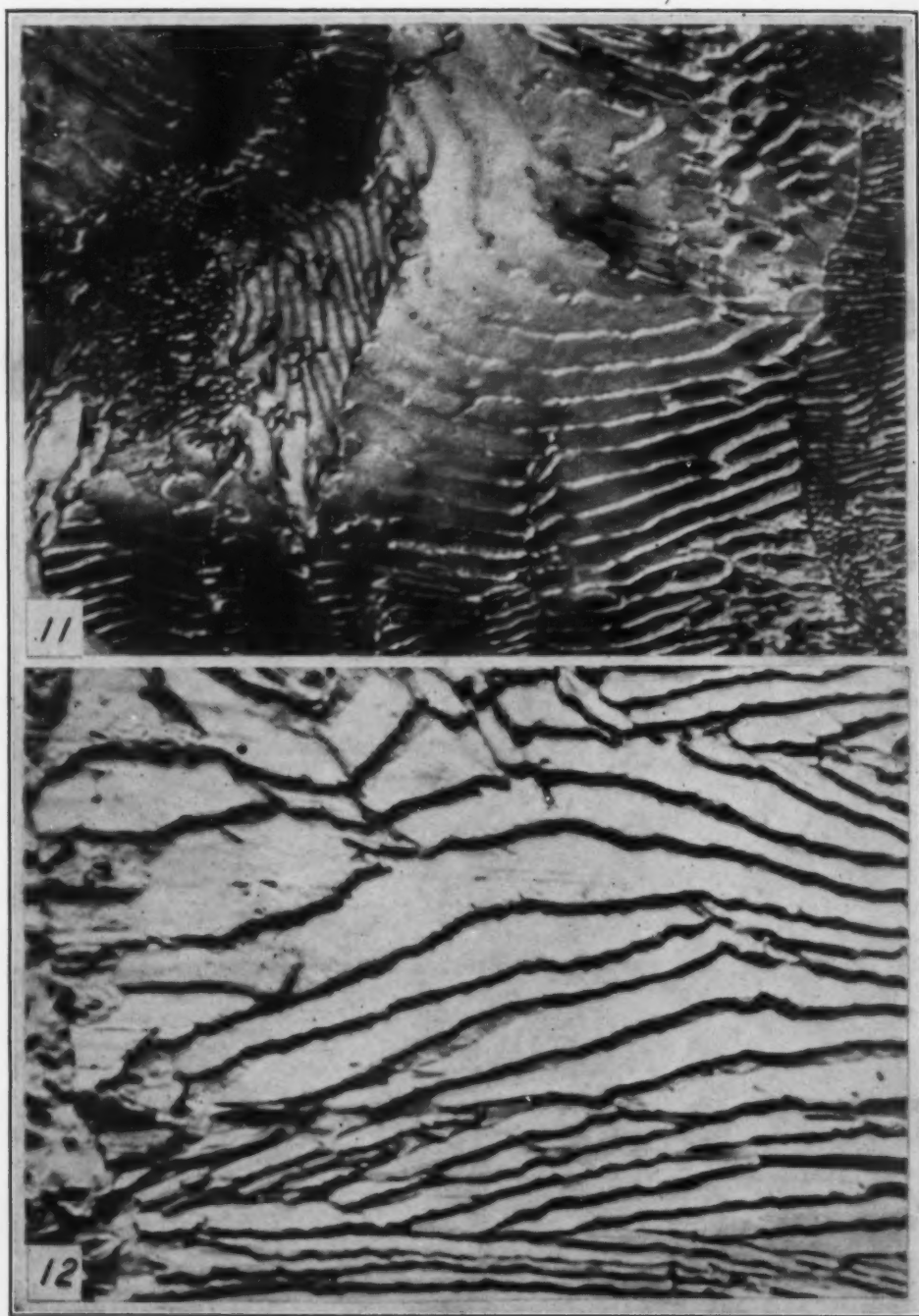


Fig. 11—Photomicrograph of Rail No. 3,  $\Phi = 56$  Degrees for Grain to the Right, 80 Degrees in Center, and 62 Degrees Grain to the Left Center.  $\times 2500$ . Fig. 12—Photomicrograph of Rail No. 10,  $\Phi = 79$  Degrees.  $\times 2500$ .

The effect of deep etching is shown in Fig. 18. The lamellae have been deeply eroded and are seen in definite steps, the appearance of which is not unlike that of weathered sedimentary rocks.

It is evident that the pearlite may be quite coarse or fine independent of the carbon content. Therefore, its characteristics will be dependent only on the velocity of cooling through  $A_{r1}$ . Thus the nature of the pearlite in eutectoid rail steel will be indicative of the finishing temperature and the rate of cooling on the beds.

#### BRINELL HARDNESS TESTS

The Brinell tests were made with a 10-millimeter ball the diameter of which varied  $\pm 0.0019$  millimeter. The length of time allowed for the application of the 3000 kilogram load was 30 seconds.

The Brinell hardness numbers were calculated by the following:

$$\text{Brinell Hardness} = \frac{P}{\pi D \left( \frac{D}{2} - \sqrt{\frac{D^2}{4} - \frac{d^2}{4}} \right)}$$

Where  $P$  is load applied,  $D$  is diameter of ball,  $d$  is diameter of impression.

An average of seven impressions was taken as an accurate value of the Brinell hardness of each specimen, i. e., seven for each tensile specimen; fourteen for each rail head.

Early in the investigation attempts were made to correlate the grain size with the Brinell hardness. However the average grain diameters measured on photomicrographs at 500 diameters of magnification bore no relation to the Brinell hardness numbers. It was found that the Brinell numbers might vary 40 points without appreciable change in average grain diameter. However, it was found that a definite relation existed between the distance between the lamellae,  $\Delta_o$ , and the Brinell hardness. Table III shows the specimens arranged in order of increasing  $\Delta_o$ . As  $\Delta_o$  increased the Brinell hardness, with few exceptions, decreased.

The product of the Brinell hardness and  $\Delta_o$  is seen to be of fairly constant value. Fig. 20 has been plotted from the data in Table III. The average curve assumes the form of a rectangular hyperbola referred to its asymptotes as axes. The equation for this curve is:

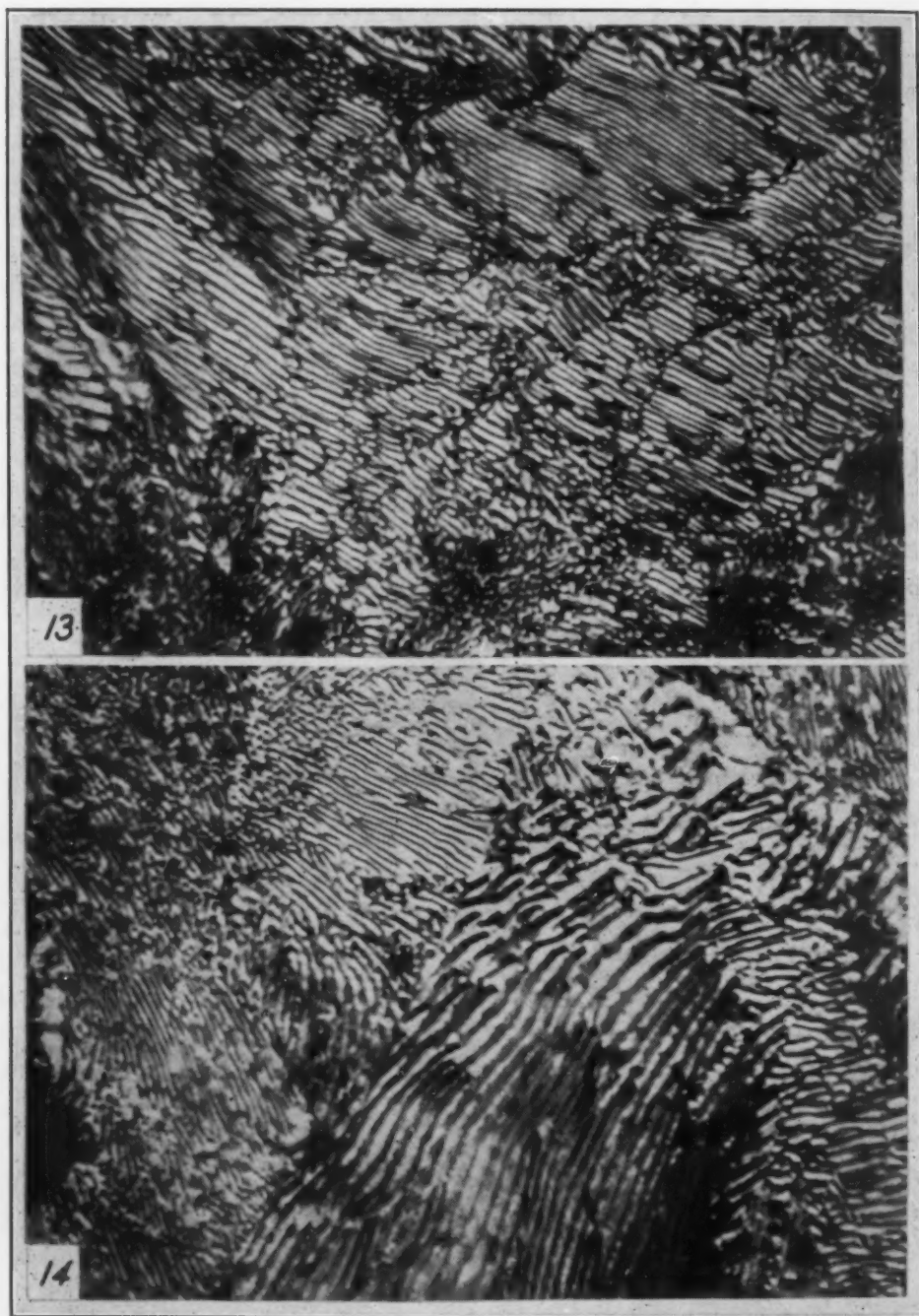


Fig. 13—Photomicrograph of Rail No. 5,  $\Phi = 0$  Degrees for Grain in Upper Corner,  $\Phi = 49$  Degrees for the Balance of the Grain.  $\Delta_o = 0.266 \mu$ .  $\times 2500$ . Fig. 14—Photomicrograph of Rail No. 8,  $\Phi = 67$  Degrees for Large Dark Lamellæ, and 25 Degrees for Bright Upper Lamellæ.  $\times 2500$ .

$$\text{No. 1—Brinell hardness} \times \Delta_0 = 79.59.$$

$\Delta_0$  in microns.

The observations in Table III were checked five times during the investigation. Consequently it is probable that the constant evaluated from this set of observations is fairly accurate. This expression is of course applicable to steels of eutectoid composition only. By the use of this expression, if one of the unknowns be determined, the other may be found with considerable accuracy.

A word of caution in interpreting the Brinell tests on rails is almost an inexcusable digression at this point. But in the author's opinion it is necessary. As many attempts have been made to correlate the Brinell hardness with the resistance of the rails to abrasion. Our experience has shown that the Brinell hardness has little or no relation to the abrasive wear of rails. It is possible that the Brinell hardness may be indicative of the resistance of a rail to plastic flow or pressure. However, the Brinell hardness tests made here are not used. For such tests it is common practice to use a 19-millimeter ball with a load of 50,000 kilograms.

#### TENSILE PROPERTIES

The proportional limits in the tensile tests were determined by the means of an extensometer reading to 1/1000 of an inch. Table IV shows the results of these tests and includes the Brinell hardness. The specimens are arranged in order of increasing Brinell hardness. With slight variations the Brinell hardness increases with the ultimate tensile strength. The relation between the ultimate tensile strength and the Brinell hardness is expressed by:

$$\text{No. 2—Ultimate Tensile Strength, Pounds per square Inch} = 490 \times \text{Brinell hardness.}$$

The ultimate tensile strength divided by the Brinell hardness of each of the specimens gave quotients of somewhat the same magnitude. The arithmetical mean of these values was used for the value of the constant in expression No. 2. The values for the ultimate tensile strength calculated by this equation are included in Table IV.



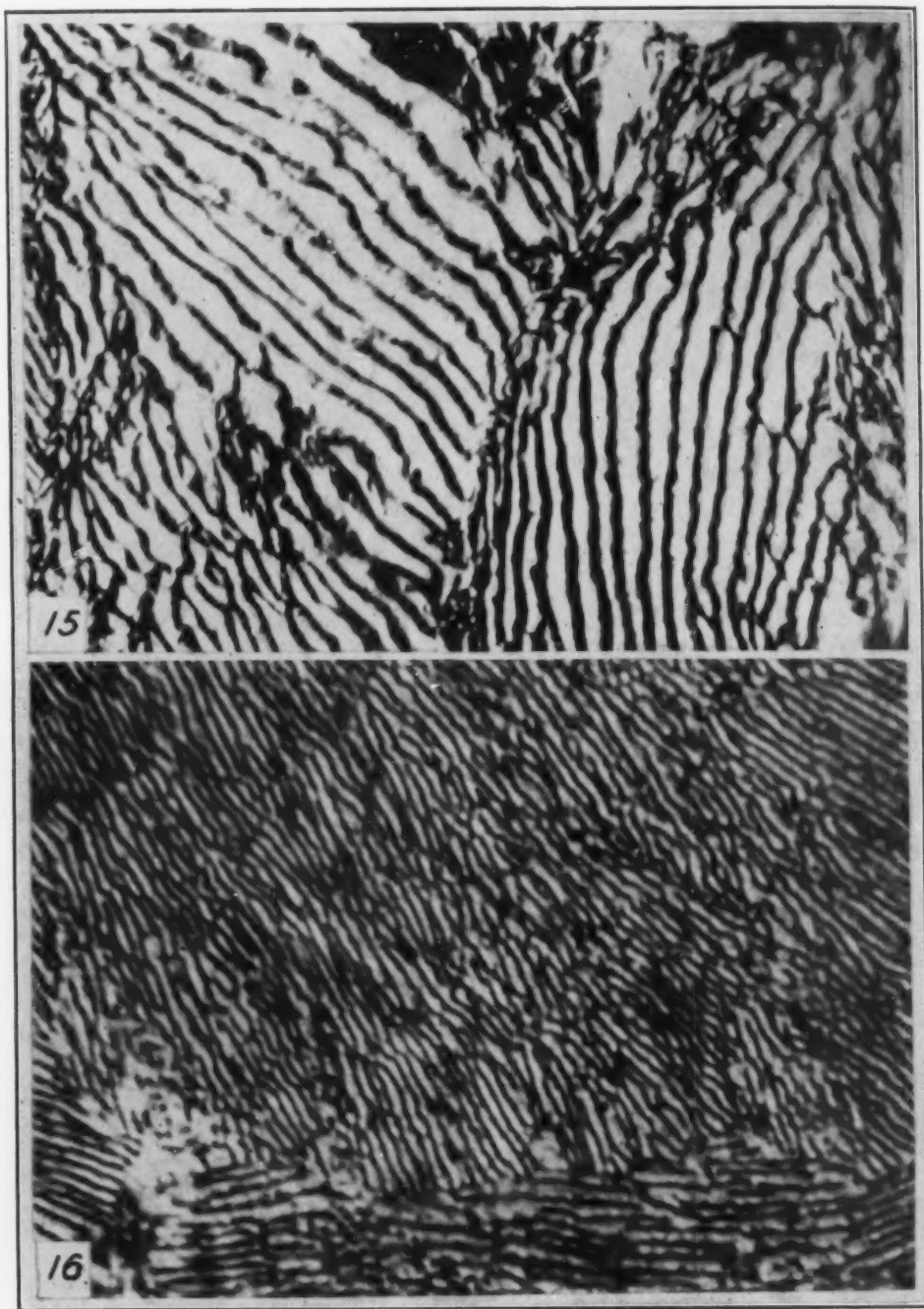


Fig. 15—Photomicrograph of Rail No. 7;  $\Phi = 69$  Degrees.  $\times 3750$ . Fig. 16—Photomicrograph of Rail No. 7,  $\Phi = 40$  Degrees.  $\times 3750$ .



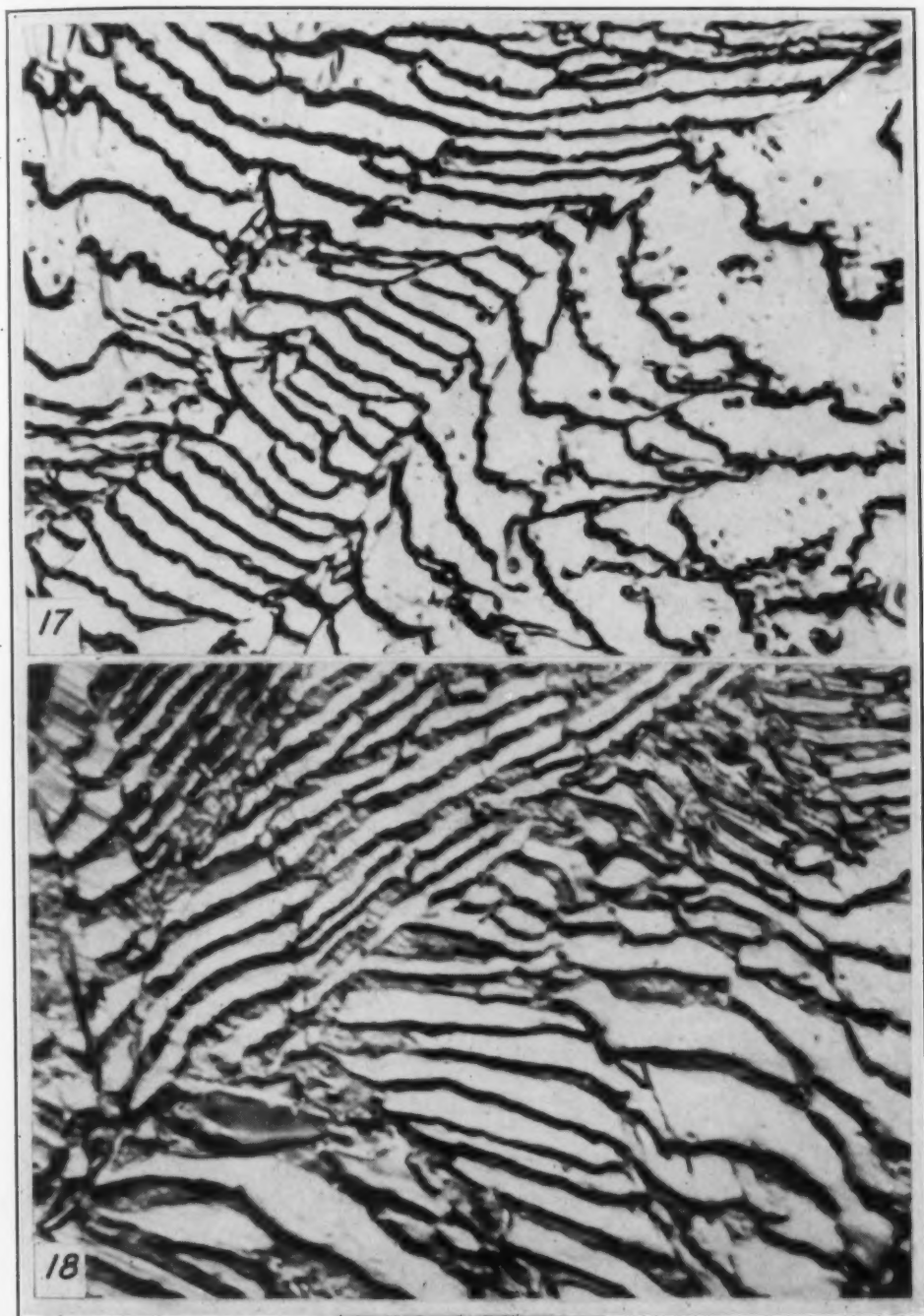


Fig. 17--Photomicrograph of Rail No. 13,  $\phi = 84$  Degrees for Largest Lamella, and Varies from 71 Degrees in Middle to 79 Degrees in Upper Left.  $\times 2500$ . Fig. 18--Photomicrograph of Rail No. 13, Specimen Deeply Etched,  $\phi = 73$  Degrees Upper Left Center and 80 Degrees Lower Right.  $\times 2500$ .

A summation of the properties of pearlite in eutectoid rail steels is given in Table V. The specimens are arranged in order of increasing values for  $\Delta_o$ . As the pearlite becomes coarser, the actual distance between the lamellae increases while the Brinell

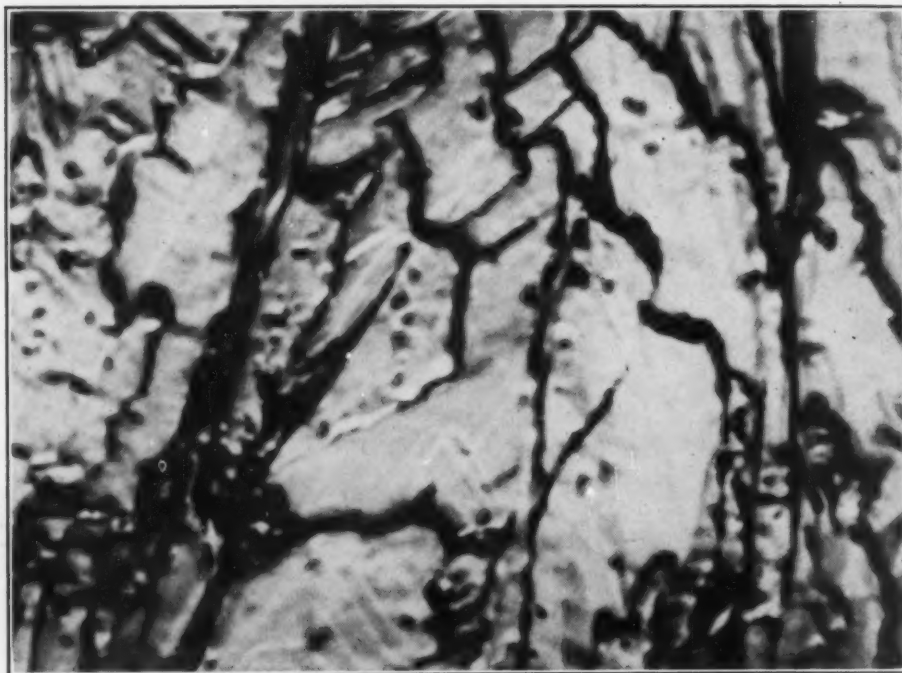


Fig. 19—Photomicrograph of Rail No. 1,  $\Phi = 83$  Degrees. Specimen Deeply Etched.  $\times 5000$ .

hardness and ultimate tensile strength decrease. The relation between  $\Delta_o$  and the ultimate tensile strength is given by:

$$\text{No. 3—Ultimate Tensile Strength, Pounds per square inch} = \frac{39000}{\Delta_o \text{ in microns.}}$$

Which has been derived from equations No. 1 and No. 2. The values for the ultimate tensile strength calculated by this expression are included in Table V.

This information demonstrates the great effects produced merely by the arrangement of the cementite plates within a steel of eutectoid composition. The specific properties of the ferrite and cementite may be altered by the presence of other elements, but under such conditions is it not reasonable to expect to find the properties of the pearlitic aggregate altered. It may be seen from

Table IV  
Relation of Brinell Hardness to Tensile Properties

Rail Number	Brinell Hardness	Proportional Limit Lbs./In. <sup>2</sup>	Ultimate Strength Actual Lbs./In. <sup>2</sup>	Ultimate Strength Calculated From Equa. <sup>2</sup> Lbs./In. <sup>2</sup>	Elong. In 2" %	Red. of Area %
13	215	53,850	105,300	105,350	10.5	13.0
10	232	64,400	115,300	113,700	8.5	18.7
11	235	52,100	113,500	115,000	11.0	14.5
1	239	66,000	118,500	117,000	15.5	23.0
12	241	68,400	117,800	118,000	10.0	17.5
2	257	63,200	124,300	125,900	10.0	19.0
3	270	55,900	134,400	132,300	7.0	14.0
7	283	61,200	138,900	138,600	9.0	16.0
8	285	65,400	138,700	139,600	12.0	21.5
4	286	67,100	137,600	140,000	13.5	22.5
6	288	69,000	138,500	141,100	11.0	13.4
9	293	68,300	144,400	143,500	15.0	25.0
5	302	72,400	150,700	147,900	9.0	18.5

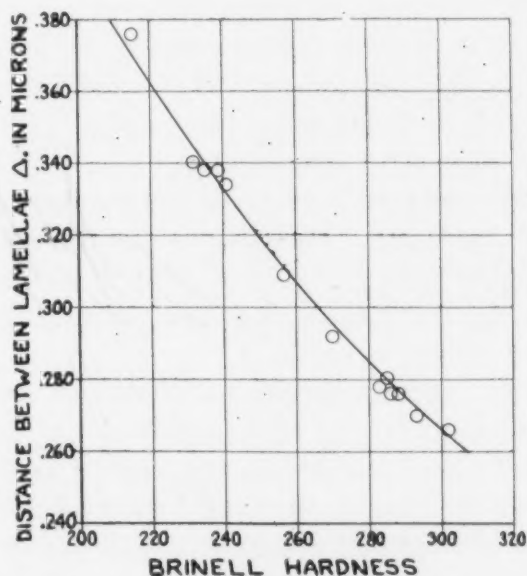


Fig. 20—Curve Showing the Relation Between the Brinell Hardness and  $\Delta_0$ , the Distance Between the Lamellae Applicable to Eutectoid Steels Only.

Table I that the manganese content varies from 0.70 to 0.81 per cent. However, it was found that this element with this variation had no effect in changing the relation between the Brinell hardness and  $\Delta_0$ .

#### SUMMARY

The apparent coarseness or fineness of the pearlite within a

**Table V**  
**Corresponding Values for Distance Between Lamellae  $\Delta_0$ , Brinell Hardness and Ultimate Strength**

Rail Number	$\Delta_0$ in Microns	Brinell Hardness	Actual Ultimate Strength Lbs./In. <sup>2</sup>	Ultimate Strength Calculated From Equ. #3 Lbs./In. <sup>2</sup>
5	0.266	302	150,700	146,500
9	0.270	293	144,400	144,400
6	0.276	288	138,500	141,300
4	0.276	286	137,600	141,300
7	0.278	283	138,900	140,200
8	0.280	285	138,700	139,200
3	0.292	270	134,400	133,500
2	0.309	257	124,300	126,200
12	0.334	241	117,800	116,800
1	0.338	239	118,500	115,300
11	0.338	235	113,500	115,300
10	0.340	232	115,300	114,700
13	0.376	215	105,300	103,700

single specimen is explained by the space relations of the lamellae and is amplified by the photomicrographs.

The pearlite may be fine or coarse quite independent of the carbon content. The character of the pearlite will depend merely on the velocity of cooling through  $Ar_1$ .

No relation was found between the average grain diameter and the Brinell hardness. However, a definite relation was seen to exist between the Brinell hardness and the distance between the lamellae,  $\Delta_0$ . As the Brinell hardness increased  $\Delta_0$  decreased, as expressed by equation No. 1.

The relations between the Brinell hardness and the ultimate tensile strength, and  $\Delta_0$  the ultimate tensile strength were also evaluated mathematically, given in equations 2 and 3 respectively. As was expected the Brinell hardness increased with the ultimate tensile strength but both decreased with increasing values of  $\Delta_0$ .

### DISCUSSION

**Written Discussion:** By J. S. Vanick, International Nickel Co., Bayonne, N. J.

The paper by O. V. Greene is the kind where the only comment possible is that of endorsement of a piece of work well done. Those of us involved in metallographic studies are aware of the vast difference in fineness of the constituent labeled as pearlite. In resolving pearlite into its components, the cementite and ferrite lamellae, we are confronted with a considerable variation in the size of these components, and usually express our observations by



describing the eutectoid fine or coarse, thus conveying some degree of difference. In our laboratory this variation has been subjected to some standardization, and in my private case a eutectoid structure, the lamellae of which could be resolved at 500 diameters, became eligible for classification as pearlite. Obviously, the optical system might be specified as well, to establish the standard more definitely. It is encouraging that Mr. Greene's work was carried into a range of high magnification well beyond the limits of routine laboratory work. When the optical system fails to resolve the structure the fineness of pearlite can become so infinite as to extinguish itself in the neighboring phase, sorbite. In view of this, the genealogy of the pearlite lamellae must originate in a nuclear or misshaped embryonic form and grow as the environment favors it into the sturdy, platy pearlite which is often resolvable at 100 diameters, or even less. It is not uncommon, especially in material like cast iron where the structure is a product of composition and cooling rate, to find pearlite and sorbite adjoining each other; the fineness of the former blending into the latter. In steels, the structure, as Mr. Greene points out, is a function of the cooling rate, and where the latter is difficult to control alloying becomes necessary, to introduce the proper impediment to the development of a beautiful, but frequently, practically undesirable structure. In the case of cast iron, where control over the cooling rate is even less readily applicable, the need for alloying is more acutely felt. While I agree with the author that the character of the pearlite is a function of the cooling rate, it is a fact that in plain or alloyed cast irons, or in alloyed steels, pearlite is itself a variable, and it becomes necessary to specify the composition as well as cooling rate for the material. These two factors might be combined into a single one and the result be defined as a function of the "diffusivity." This suggestion is offered merely for the purpose of enlarging the scope of the author's paper so that other workers may find it useful to consider the vast field open to them in a study of pearlite and its relation to physical properties.

#### Oral Discussion

GEORGE F. COMSTOCK: I would like to ask the author if he found any sorbite in these rails. In all steel rails that I have examined we find sorbite as well as pearlite; and if there is sorbite present, I would like to ask how the delta sub-zero was determined. That puzzled me very much in reading the paper.

O. V. GREENE: Apparently the mill from which most of these rails were received finished very hot, and we could not find any sorbite. In some of the specimens we had a slight amount of granular pearlite in spots, but it was not sorbitic pearlite. That is, the pearlites were fairly coarse in all the rails we examined. The mills from which all these rails were taken are noted for finishing quite hot, consequently, we found no sorbite.

ROBERT JOB: Mr. Chairman, I should like to ask whether Mr. Greene has found any definite relation between structure and durability in rail steel. It opens up the broad question, of course, as to the practical application of the microscopic work and that is a matter which we all know has been developed very largely within recent years, and I think we will all agree that

it offers a tremendous opportunity for development. I should be very much interested to have Mr. Greene's comments as to just what relationship he has found between the structure which he has determined and the durability of the steel in service.

#### Author's Reply to Discussion

O. V. GREENE: In reply to Mr. Job's question, we have attempted to see if there was any definite relation between the microstructure of the pearlite and the abrasion on rails. However, to date the information we have obtained has been so conflicting that I feel it should not be given. Actually, little connection could be found between the two. Incidentally quite a definite relationship was found between Brinell hardness numbers, made at 100,000 pounds applied load with a 19 millimeter ball on top of the heads, and the resistance against plastic flow. However no connection could be found between the microstructure and the flow of metal in the heads in service. Consequently that field is still very open. I thank Mr. Job for the idea and the question. I think it perhaps can be carried farther and we undoubtedly can get some very valuable information.

In the written discussion by Mr. Vanick, the question of the resolution of the pearlite grains was raised. The method of examination used in finding grains in which the lamellae are normal to the section cut is as follows: The sections are examined under increasing magnification; grains that are not resolved at 500 diameters, for instance, are examined at higher magnifications. At 750 diameters some of the grains not resolved at 500 diameters are now well delineated. The unresolved grains are examined at 1000 diameters, and any remaining unresolved grains are again examined at increasing magnification. The pearlite in all of the steels investigated was resolved at 2500 diameters. At 2500 diameters the grains containing the finest lamellae were each measured and an average figure for delta sub-zero taken as representative of the sample. As I told Mr. Comstock a bit earlier, there was some granularity present in the specimens but in no case could it be called sorbite.

The question was also raised concerning the effect of alloys. I think it is generally understood that vanadium, for instance, will replace the iron in the cementite, until at 5 per cent. vanadium all the iron has been replaced, and vanadium carbide is present instead of iron carbide in the eutectoid. Both molybdenum and chromium are known to lower the carbon content which the eutectoid forms.

## CLOUDBURST PROCESS FOR HARDNESS TESTING AND HARDENING

BY EDWARD G. HERBERT

### *Abstract*

*It has been found that if a piece of steel be placed in a jet of hard steel balls moving at a high velocity, the balls will rebound from a hard steel surface leaving it unaffected but will roughen a soft surface. By this means the exact shape and location of soft areas may be determined, whereas by present hardness testing methods only one or a few selected spots are tested.*

*It has also been found that it is possible to superharden a steel surface artificially by bombarding the surface with hard steel balls. It is believed that this process will be used to superharden the teeth of gears and other parts where extreme hardness is necessary.*

THE cloudburst process consists essentially in bombarding an object to be tested for hardness or increased in hardness with large quantities of hard steel balls, which are caused to strike the object with predetermined and controlled velocity.

In practice the most convenient method of causing balls to move with controlled velocity is that of allowing them to fall from a known height, and this method is adopted in the simplest form of cloudburst machine, shown in Fig. 1. The process is carried on in a rubber-lined chamber from the center of which rises a pipe having ports near the top. Steel balls, to the number of 20,000 or more in the smallest machines, and usually of 3 or 5 millimeters in diameter, are placed in a hopper surrounding the pipe. The hopper being raised to the top of the pipe the balls pour through the ports and fall down the pipe into the chamber.

The movement of the balls inside the pipe is first controlled by a regulating valve consisting of a series of radial grids suspended just below the ports. These serve to impede, and, by a suitable adjustment to regulate, the rate of flow of the balls. Having passed the regulating valve the balls fall onto a perforated piston suspended inside the pipe, and capable of being set at any desired

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A paper presented before the tenth annual convention of the society held in Philadelphia, October 8 to 12, 1928. The author, Edward G. Herbert, member of the society is connected with Messrs. Edward G. Herbert, Ltd., of Manchester, England. Manuscript received June 29, 1928.

height above the work table in the chamber. The piston arrests the falling balls, but permits them to start afresh by passing through its perforations. The effective height of fall, and therefore the effective velocity with which the balls strike the work is thus determined by the height of the piston within the pipe. This is set by the slider and scale seen in the illustration, and may be from 0 to 4 meters. (13 feet)

The jet of balls thus produced is circular in section and about 2 inches in diameter. The work is placed on a table suspended within the chamber by four rods provided with ball joints. An indicator on the top of the chamber shows what part of the table is actually in the direct ball jet at any moment, and renders it easy to distribute the bombardment evenly over the work on the table or to concentrate it on a particular area as may be desired.

A rotary motion is provided for treating gears and other cylindrical objects and is driven at a slow speed by a handle and chain gearing.

In the machine shown in Fig. 1 a shower of 20,000 balls can be produced in from 10 to 30 seconds, and can be concentrated on a 2-inch circle or distributed over a table 8 inches square.

Various styles of cloudburst machines are under construction, in which a continuous flow of balls is maintained by bucket elevators or other means. In the larger machines the jet of balls takes the form of a thin sheet, the balls being caused to fall into the treating chamber through a long narrow slot in the adjustable piston. The work table is caused to reciprocate beneath the flat jet of balls so that every part of its surface is acted on at each pass. For treating work in large quantities or articles of great length it is proposed to employ a conveyor to carry the work through the treating chamber.

#### THE CLOUDBURST HARDNESS TESTS

*The Elimination Test*—If a piece of steel be placed in a jet of hard steel balls moving with a certain velocity, the balls will rebound leaving the surface unaffected if the steel be hard, but will roughen the surface by countless indentations if the steel be soft, while in the case of a piece of steel which is partly hard and partly soft, (for instance a badly case hardened article on which there are soft spots) the exact shape and location of the soft areas



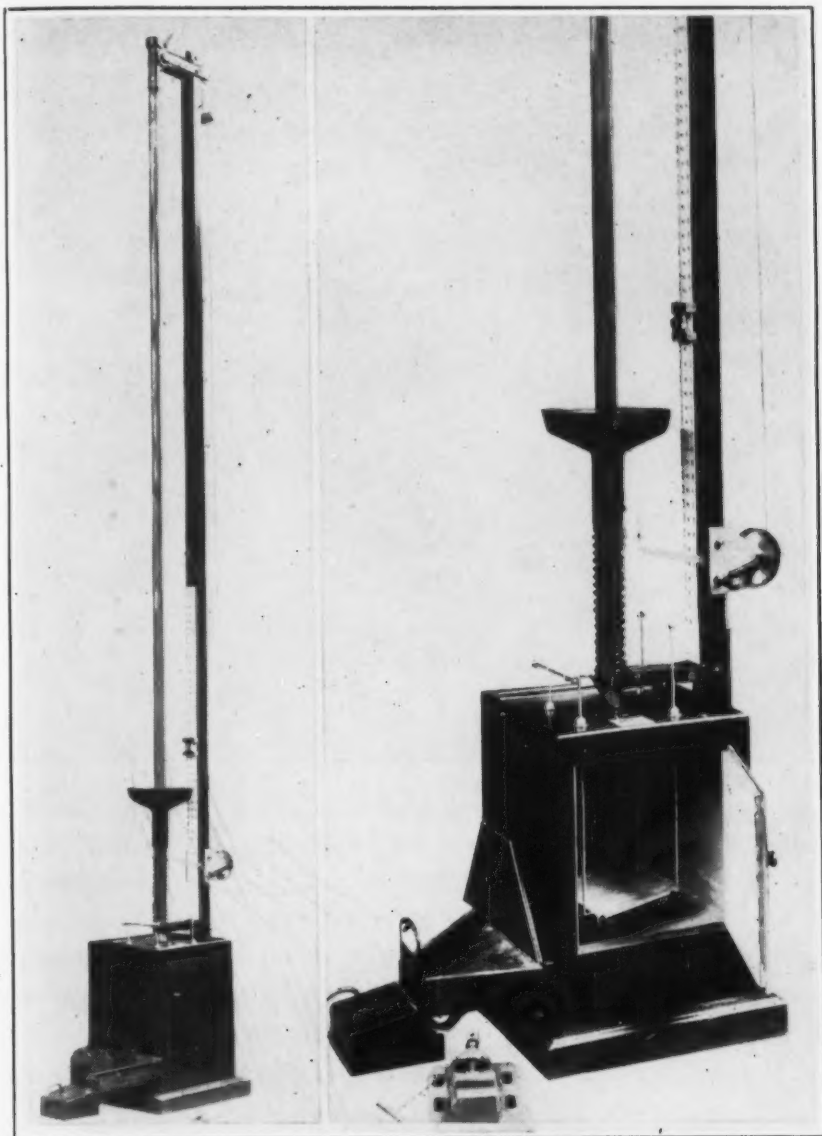


Fig. 1—The 8-Inch Cloudburst Machine. Maximum Height of Fall 4 Meters.

will be revealed by their indented appearance, while the hard areas will be unmarked. By selecting a suitable ball velocity a surface of any desired hardness may be left unmarked, while any surface of less than the specified hardness will be rendered easy of detection by its indented appearance.

Most of the hardness inspection which is at present carried out has for its sole object the detection of inadequate hardness. The advantages of the cloudburst elimination test are that it en-



Fig. 2.—Photograph of the Projection Microscope.

ables large numbers of articles to be tested simultaneously, that it tests the whole surface instead of only one or a few selected spots, and that articles of correct hardness are unmarked by the test, while any soft areas which exist are revealed in size, shape and location. With a little practice it is easy to "read" a specimen which has been so tested, judging the degree of hardness or softness of the whole surface by simple inspection without any measurement.

*The Measurement Test*—In cases where it is necessary to make an accurate measurement of hardness for the purpose of controlling it between limits, the procedure is similar, but a smaller number of impacts are given so that the impressions do not overlap and their diameter can be measured.

To facilitate measurement of impressions the special projection microscope shown in Fig. 2 is provided. The work, after having been placed in the cloudburst test for a few seconds, is held against the hard steel nose piece at the bottom of the projector, and the operator, looking through the window at the front,

sees a brilliant magnified image of the surface dotted with impressions, projected on a white graduated screen 8 inches in diameter.

The optical system consists of a series of lenses, mirrors, and prisms and a powerful projector lamp. One of the mirrors can be tilted by means of a screw so as to give a slight vertical movement to the image on the screen, and in this way the edge of any

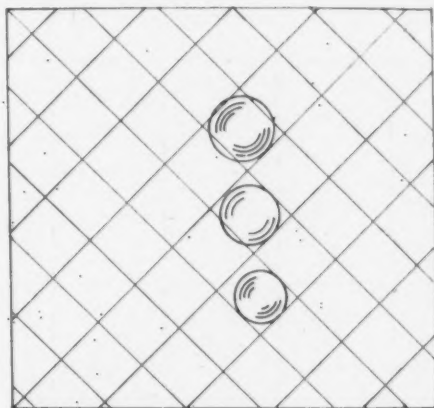


Fig. 3—Limit Screen for Projection Microscope.

impression can be brought into coincidence with one of the graduation lines with which the screen is ruled. The standard screen has horizontal graduation lines corresponding to tenths of a millimeter.

When it is desired to work between predetermined limits of hardness the standard screen is replaced by a "limit screen" Fig. 3 ruled with two sets of graduation lines inclined at 45 degrees to the horizontal. Thus the screen is divided into rectangles whose sides correspond in length with the limiting diameters of impressions. By tilting the mirror any impression can be brought to coincide with either series of graduations, and if the hardness is correct the impressions will be not less in diameter than the shorter nor greater than the longer sides of the rectangles. Thus all fine graduations are dispensed with, however close the limit adopted.

The cloudburst test does not add to the too numerous scales of hardness which already exist. Conversion can be readily made from the diameter of cloudburst impressions to Brinell, Rockwell, Shore, Pendulum, or any other hardness scale.

## SUPERHARDENING

In a paper by the author (The Work-hardening of Steel by Abrasion, *Journal, Iron and Steel Institute* 1928) it was shown that articles of hard steel such as gears and cams of automobiles, when subjected to severe service conditions are "superhardened." A superficial layer of intense hardness is produced on any hard steel surface which has been worn under conditions so severe as to

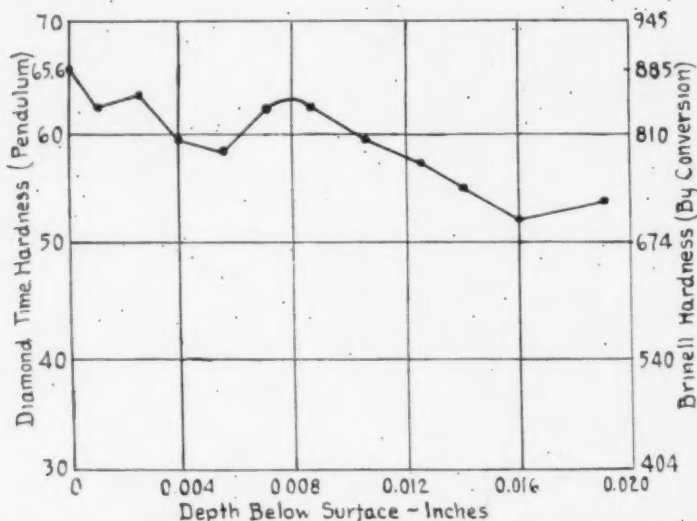


Fig. 4—Hardness Gradient Below Surface of Large Roller Bearing, Superhardened by Overloading.

cause plastic deformation or flow of the steel. Many examples of this were given in the paper referred to, the hardness induced by wear on the surface of cams from automobiles rising in some cases to the equivalent of 1160 Brinell.

Fig. 4 shows the hardness gradient below the surface of a large roller bearing outer ring which had been overloaded in service. The hardness is given in terms of diamond-time hardness, and also in the equivalent Brinell hardness obtained by using the conversion factor 13.5. Tests were made with the pendulum and 1-milimeter diamond on a small area of the worn surface, the average surface hardness being 65.6 (885 Brinell). The surface was then removed to a depth of 0.0012 inch by hand rubbing with a carborundum hone and a further series of pendulum tests was made on the surface so produced, and this was continued until the original hardness was reached. As the pendulum weighs only 4

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kilograms and the depth of the impression made on hard steel is inappreciable, it is possible by this method to obtain a hardness gradient by steps of 0.0005 inch or less.

It is evident that a considerable degree of superhardening had been caused by overloading the bearing, and that it extended to a depth of 0.016 inch below the surface. The maximum hardness occurred on the surface, and immediately below it is a curious depression in the hardness curve, which appears to be characteristic

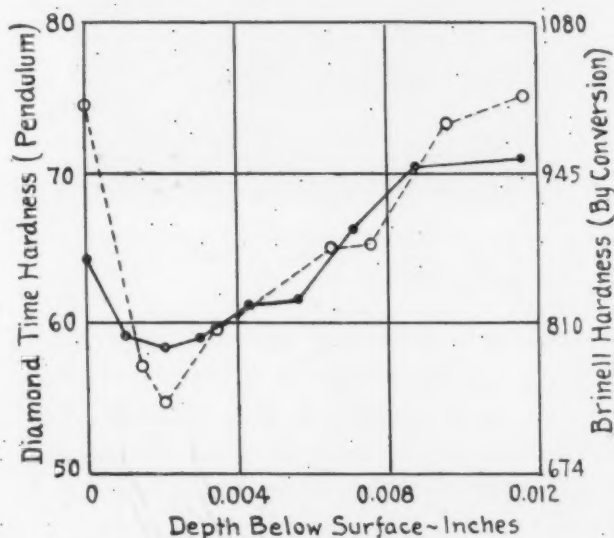


Fig. 5—Hardness Gradient Below Surface of Two 10-Millimeter Steel Balls Superhardened by Overloading.

of superhardening induced by a rolling process. A similar drop in hardness immediately below the surface is seen in Fig. 5 which shows the hardness gradients below the surface of two 10-millimeter steel balls which had been superhardened by rolling under excessive load. In each case the surface hardness is high, but does not approach the maximum superhardening capacity of the steel. This can be measured by the time workhardening test made with the pendulum and 1-millimeter diamond ball. A time test is first made to find the original hardness. The original circular impression is then rolled out into an elongated form Fig. 6 and the steel superhardened, by tilting the pendulum, and a further time test is made on the rolled surface. By alternately testing for hardness and superhardening by rolling a series of hardness numbers is obtained. These invariably increase to a maximum and decline, as is also the case with soft metals. A time workhardening test

made on an unworn surface of the bearing ring Fig. 4 gave the following results:

Passes of diamond .....	0	2	4	6	8	10
Diamond time hardness ....	53.6	77.0	77.1	79.3	81.0	79.1

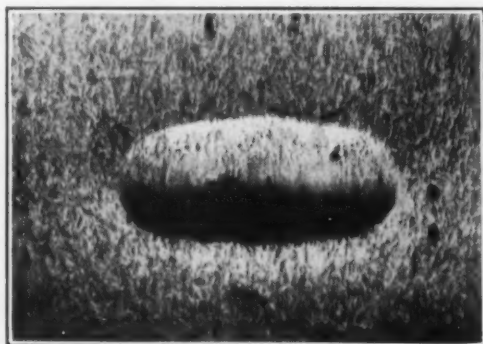


Fig. 6—Impression in Hard Steel Made by Pendulum and 1-Millimeter Diamond in Time Workhardening Test.  $\times 100$ .

Thus the original hardness of the steel was 53.6 (723 Brinell) and the maximum superhardness induced by rolling 8 times was 81 (1094 Brinell), as compared with maximum superhardness induced by overloading the bearing 65.6 (885 Brinell).

From these and many other examples it is clear that the superhardening of hard steel is what may be termed a "natural" process occurring every day in ball bearings, automobiles, and any other mechanisms containing hard steel parts which are liable to be subjected to action so severe as to cause displacement or flow of the surface metal. It occurs for instance whenever a hard steel tool is struck with a hammer.

In the cloudburst superhardening process the same effect is produced artificially by bombarding the hard steel surface with hard steel balls. Balls 3 millimeters in diameter are usually employed and the initial ball velocity is determined by the original hardness of the article to be treated. If too high a ball velocity were employed at first, the surface would be roughened by indentations. The initial velocity must be such that the surface layer of steel is slightly displaced but not sufficiently to roughen it. When the surface has been completely covered by light blows it is found to be encased in a thin superhardened layer, which is capable of resisting indentation by balls striking it with increased velocity but is thereby increased in hardness and in thickness. By progres-

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sively increasing the ball velocity a smooth surface of intense hardness is produced, the hardness diminishing in an even gradient as shown in Fig. 7 which is the hardness gradient below the surface of a high carbon steel hardened by quenching and superhardened by bombardment at increasing velocity with 3-millimeter balls. Fig. 8 is the gradient of a heat treated nickel-chromium steel

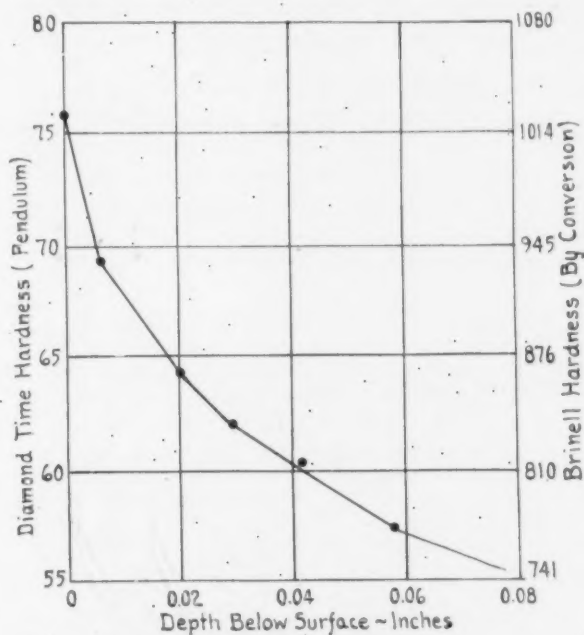


Fig. 7—Hardness Gradient Below Surface of 1.25 Per Cent Carbon Steel Hardened by Heat Treatment and Super-Hardened by Cloudburst.

after superhardening. The final and maximum ball velocity which can be employed without roughening the surface depends upon the superhardening capacity of the steel, which is easily measured by the pendulum time-workhardening test.

#### CLOUDBURST WORKHARDENING

Just as the cloudburst hardness test on hard steel can be prolonged to superharden the surface, so can the same method be applied to induce a workhardened layer on soft steels and other metals. The principle is the same,—the initial ball velocity must be adjusted to the original hardness of the surface so as not to roughen it, and when the first stage of low velocity bombardment has been completed the ball velocity is increased progressively until

it reaches a maximum (depending on the workhardening capacity of the metal) which is sufficient to induce maximum hardness but insufficient to roughen the surface. An actual example will best illustrate the process.

In Fig. 9 is shown the result of workhardening 14 per cent manganese steel by the cloudburst process. The steel was first tested with the pendulum time-workhardening test and was found to have an original time hardness 27.6 (276 Brinell) and a maximum induced hardness of 71 (710 Brinell) occurring after 12 passes of the pendulum ball, as shown in the upper curve of

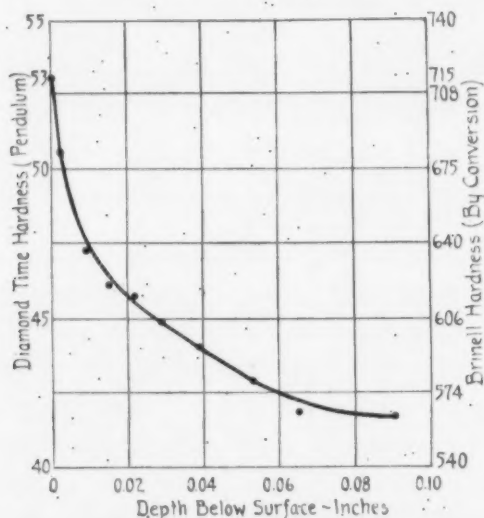


Fig. 8—Hardness Gradient Below Surface of Nickel-Chromium Steel Heat Treated and Super-Hardened by Cloudburst.

Fig. 9. The specimen was then placed in the cloudburst of 3-millimeter balls falling from a height of 0.5 meter, whereby the hardness was increased to 40 (400 Brinell). The height of fall was then increased to 1 meter and the bombardment was resumed. By progressively increasing the height of fall to 4 meters the final surface hardness of 665 Brinell was induced. The hardness gradient below the surface of this specimen was measured with the pendulum as described above and is shown in Fig. 10. The hardening operation was carried out in the hand operated machine Fig. 1 and the actual time occupied was not measured. The process was completed by 360,000 blows per square inch of the surface. Workhardening and superhardening will ordinarily be carried out in machines in which a continuous flow of balls is secured by means



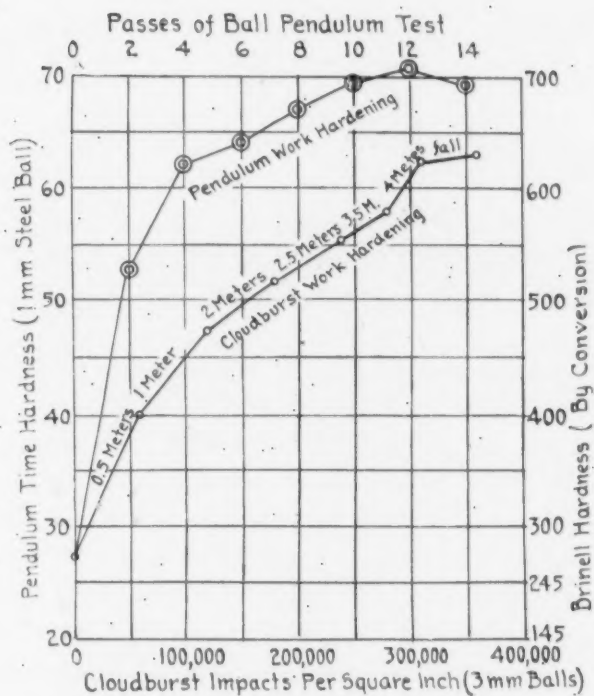


Fig. 9—Curves Showing Results of Pendulum Time Workhardening Test and Cloudburst Workhardening Process on Manganese Steel.

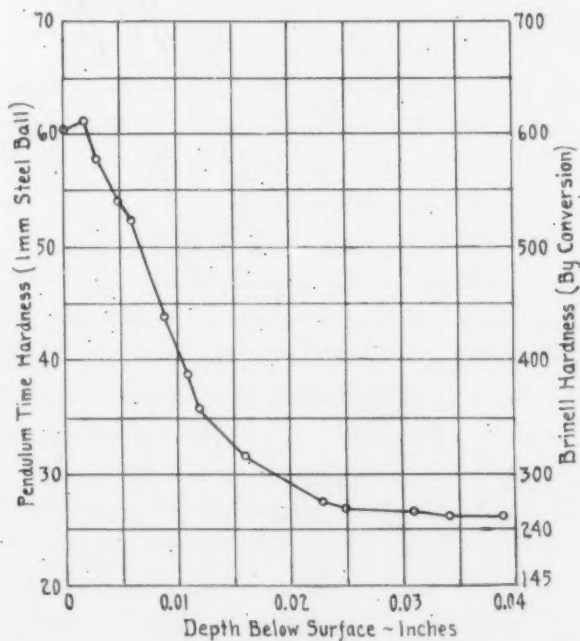


Fig. 10—Hardness Gradient Below Surface of Manganese Steel Hardened by Cloudburst.

of a ball elevator. The smallest of these machines has a work table 12 by 12 inches and is designed to give 500,000 blows per minute.

The commercial applications of the cloudburst hardening processes are numerous, and are now under development. Of these the artificial superhardening of hard steel by bombardment as a means of increasing resistance to wear is a new process whose application to the teeth of gears and so forth has yet to be worked out.

The cold working of steel in order to increase its hardness and strength is however so well known and its efficacy so well established in the production of hard drawn wire, tubes, and sheets, that no more than a brief reference is necessary. The cloudburst is a process which can be applied to surfaces of any shape, even to internal surfaces by the use of suitable deflectors,—and the extent of its useful applications can only be shown by experience. The surface hardening of rails is only one application which may be expected to assume importance. The possibilities of its application to metals other than steel are scarcely less interesting. Thus cast iron, whether in the soft state or chilled, is shown by the pendulum workhardening test to possess an abnormally high workhardening capacity, and should therefore lend itself to the cloudburst process in the surface hardening of such articles as, car wheels, chilled iron rolls, and cylinder bores. These commercial applications have yet to be developed, but it may be useful to indicate some of the metallurgical phenomena which have to be taken into account in carrying out the process.

It will have been observed that in measuring the workhardening capacity of a metal by means of the pendulum, the hardness induced by cold work invariably rises to a maximum and declines, see Fig. 9. If the cold working process is continued, the hardness commonly rises to a second maximum, which is sometimes higher, sometimes lower than the first. This phenomenon manifests itself to a marked degree in cloudburst hardening. It has not so far been observed in hard steel superhardened by the cloudburst, though the peculiar form of the hardness gradient in the overloaded roller bearing Fig. 4 suggests that some such fluctuation may be expected to occur. The phenomenon has however been very clearly indicated in the cloudburst hardening of soft steels. In

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such cases the hardness rises to a maximum which is generally lower than the pendulum maximum, and declines when the process is continued. Researches now in progress appear to be throwing light on this subject but are not sufficiently advanced for useful discussion.

Another phenomenon to which considerable attention has been given in recent years in America and England is the effect of low temperature annealing on cold-worked metals. Reference may be made to the work of Langenberg<sup>1</sup> and Grossmann and Snyder<sup>2</sup> in America and to that of Lea in England<sup>3</sup>. Fig. 11, reproduced

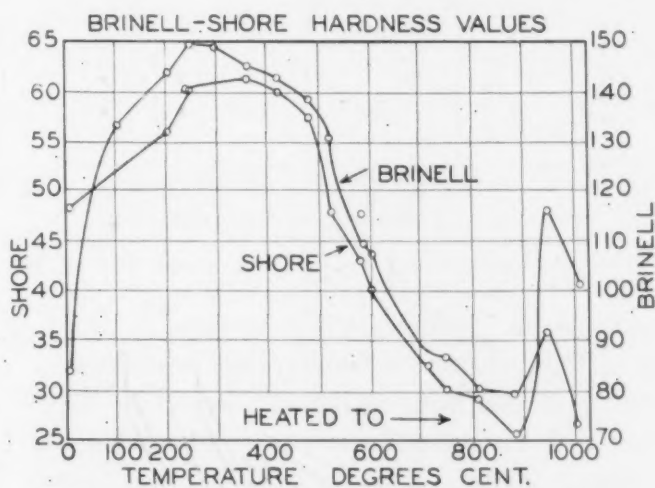


Fig. 11—Curves Showing the Effect of Annealing Temperature on Hardness of Cold-Drawn Steel Tubes. (F. C. Lea.)

by the courtesy of the editor of *Engineering*, shows the effect on hardness of annealing cold drawn tubes at various temperatures (F. C. Lea). A marked increase of hardness results from annealing the cold-worked steel at 300 degrees Cent. A similar increase of hardness has been secured by low temperature annealing of steels which have been hardened by the cloudburst process.

The effect of low temperatures on the workhardening capacity of metals has been a subject of investigation by the present author for several years. Fig. 12 shows the effect of temperature on the

<sup>1</sup>F. C. Langenberg, "Effect of Cold Working on the Strength of Hollow Cylinders." TRANSACTIONS, American Society for Steel Treating, 1925.

<sup>2</sup>M. A. Grossmann and C. C. Snyder, "Hardening by Reheating after Cold Working." TRANSACTIONS, American Society for Steel Treating, 1927.

<sup>3</sup>F. C. Lea, "The Effect of Heat Treatment on Cold-Drawn Steel Tubes." *Engineering*, Dec. 23 and 30, 1927.

workhardening capacity of a low carbon steel, (0.18 per cent carbon). The characteristic form of curve with three maxima and three minima has been found not only in steel but in pure metals and nonferrous alloys<sup>1</sup>. In steel the maximum  $P_3$  usually occurs in the neighborhood of 300 degrees Cent. and is believed to be connected with the "blue brittleness" phenomenon, and

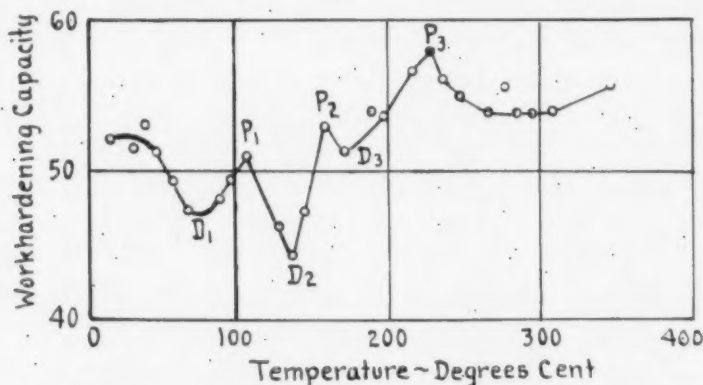


Fig. 12—Workhardening Capacity and Temperature of Low Carbon Steel.

with the increase of hardness which occurs in steel as a result of annealing at that temperature after cold working.

An instance of the application of low temperature annealing in conjunction with the cloudburst process is shown in Fig. 13. A low-carbon steel, Brinell hardness 105, was tested with the

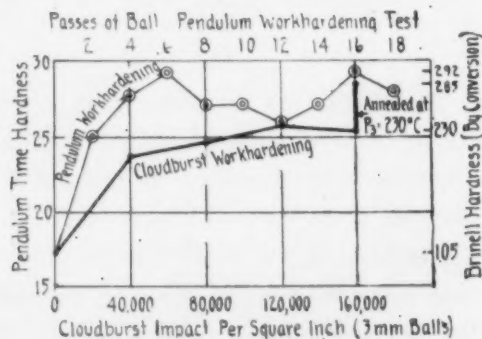


Fig. 13—Curves of Tests on Mild Steel Tested with Pendulum Time Workhardening Test, Workhardened by Cloudburst and Annealed at  $P_3$  Temperature 230 Degrees Cent.

pendulum time-workhardening test, (see upper curve Fig. 13). The hardness rose to 292 Brinell after 6 passes of the pendulum ball, declined to a minimum after 12 passes, and increased to a

<sup>1</sup>E. G. Herbert, "Work-hardening Properties of Metals." American Society of Mechanical Engineers, 1926.

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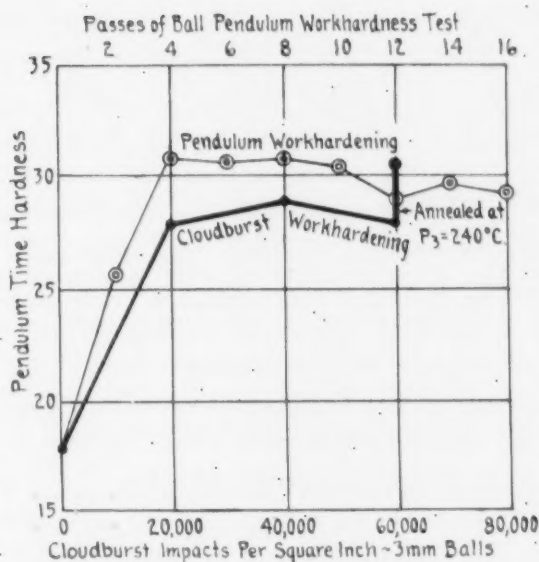


Fig. 14—Curves of Test on Sheet Brass: Tested with Pendulum Time Workhardening Test, Workhardened by Cloudburst and Annealed at  $P_3$  Temperature 240° (Degrees Cent.).

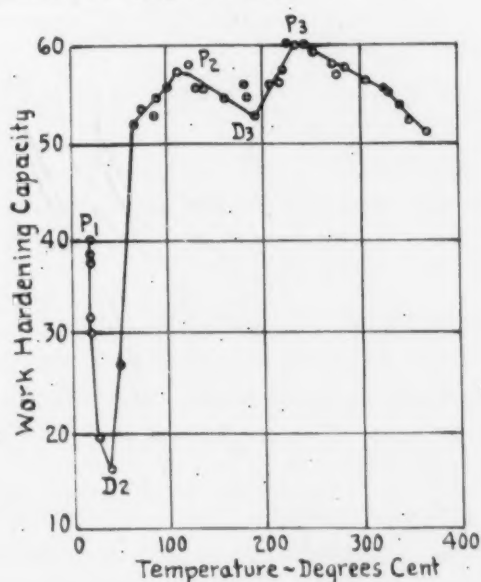


Fig. 15 — Temperature-Workhardening Curves (Pendulum) Sheet Brass.

second maximum 292 Brinell after 16 passes. The specimen was placed in the cloudburst of 3-millimeter balls, and its hardness increased to a maximum 237 Brinell after 120,000 blows per square inch (see lower curve Fig. 13) decreasing to 230 Brinell after a further 40,000 blows. The steel was then annealed for 20

minutes at the  $P_3$  temperature which was 230 degrees Cent. as shown in Fig. 12 which is the temperature-workhardening curve of the same steel. The effect of the annealing was to raise the surface hardness from 230 Brinell to 285 Brinell.

A similar application of  $P_3$  annealing to workhardened brass is shown in Fig. 14. A specimen of sheet brass was tested with the pendulum time workhardening test and gave (upper curve Fig. 14) a maximum after 4 passes, a minimum after 12 passes, and a second maximum after 14 passes. The same specimen placed in the cloudburst of 3-millimeter balls was increased from time hardness 17.7 to a maximum 28.6 after 40,000 impacts per square inch, and the hardness fell after further treatment. The specimen was then annealed for 30 minutes at the  $P_3$  temperature for this particular brass, which, as seen in Fig. 15, occurs at 240 degrees Cent. The effect of the  $P_3$  anneal was to raise the hardness from 27.8 to 31, this being almost identical with the maximum hardness obtained in the pendulum workhardening test.

#### SUMMARY AND CONCLUSION

The cloudburst process of bombardment with hard steel balls has four principal functions,—

1. The elimination of soft articles or articles having soft spots, the work being tested in quantity, all at once, and all over, and without marking surfaces of not less than a specified degree of hardness.
2. The measurement of hardness and its control between limits, the work being tested in quantity and over its whole surface.
3. The superhardening of hard steel.
4. The production of a workhardened surface on steel, cast iron, and other metals.

It is shown that the workhardened surface can be further increased in hardness by annealing at the  $P_3$  temperature, at which the workhardening capacity of the metal reaches a maximum. This temperature occurs in steels generally between 200 and 300 degrees Cent. and in brass between 200 and 250 degrees Cent.

#### DISCUSSION

**Written Discussion:** By S. N. Petrenko, Bureau of Standards, Washington, D. C.

The cloudburst process of hardness testing seems to be full of possibili-

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ties. It is possible that smaller balls and a different method of sending the balls against the surface, similar to the sand blasting process, might be used with advantage. It does not seem, however, wise to try to reduce the results of this test to other more common scales of hardness. A single Brinell, Vickers, or Rockwell test would not take much more time than making a measurement test by the cloudburst process, and it would have the advantage of being a direct test. Probably it will not be difficult to design standards for this particular test. Experience shows that attempts to carry too far the conversion of standards may lead to unnecessary complications and confusion.

**Written Discussion:** By Alvan L. Davis, Scovill Mfg. Co., Waterbury, Conn.

Like all of Mr. Herbert's contributions, the present paper is both novel and interesting. In the absence of experience with the cloudburst process, all that I can do is to ask Mr. Herbert a few questions which suggest themselves after reading his paper.

How is the conversion of cloudburst indentation made to Rockwell or to Brinell hardnesses (see last paragraph, page 81)? Referring to Fig. 4, page 82, what is considered the original hardness of the roller bearing and how is it indicated by the plot?

In the same way with the two balls shown in Fig. 5, page 83, what would be considered as the original hardness of the surface of these balls?

To what extent would the smoothly ground surface of a chilled iron roll be roughened by superhardening by the cloudburst process? If Mr. Herbert does not have the data for chilled iron perhaps he may be able to answer the same question as applying to the plain surface of hardened tool steel, after it has been superhardened by the cloudburst process.

Does chilled cast iron exhibit a similar falling-off from the maximum workhardening of the pendulum tester, to that shown in Fig. 9, page 87, for manganese steel?

I cannot close without calling attention to the interesting and beautiful exhibit of the workhardening and hardness gradient for manganese steel which is exhibited in Figs. 9 and 10. Also Fig. 14 is a very interesting exhibit as pertaining to the workhardening of sheet brass. In this connection I would like to ask what Rockwell hardness would correspond to the pendulum-time hardness of 31.

Predictions are, of course, dangerous to make, but one may hazard the judgment that the greatest field for usefulness of the cloudburst process will be the detection of soft spots on extensive surfaces, all of which are desired to be fully hardened. There is, of course, the possibility of a wide field of use for superhardening the wearing surface of hardened steel parts.

#### Oral Discussion

R. B. LEWIS\*: Taking up Captain Petrenko's specifications first, I think that Mr. Herbert has not attempted to bring out a new scale of hardness. He is endeavoring to perfect a method by which variations in hardness can be readily distinguished, and I think for that application using falling balls and

\*This paper by E. G. Herbert was presented by R. B. Lewis of the Tinius Olsen Testing Machine Co., Philadelphia.

a height which is correlated to the particular hardness that is being worked on, he has produced a very useful machine that will be applicable to the 100 per cent inspection such as we attempt to give treated material today.

Also, the Brinell, Rockwell and Shore hardness test give you a reading at some particular point or points on the object being tested for hardness. You might be fortunate in picking a soft spot, you might be fortunate in picking a hard spot, and the acceptance or rejection of that particular piece of material depends on the luck you have had in picking the spot. The cloudburst machine is an attempt to get away from this and give us a machine which will give us the average surface.

Replying to Mr. Davis' remarks, a conversion table has been developed by Mr. Herbert. This table was sent us showing what could be expected but it was not developed in time to be incorporated in this paper. However, I have the original data, a manuscript of Mr. Herbert's, showing that using the projecting microscope shown before, and five-millimeter balls falling four meters, different diameters of impression gave different Brinell numbers, corresponding to material tested. He has also made the same experiments here, using a three-millimeter ball falling four meters, and his experiments seem fairly consistent and give us a pretty good line on what we may expect in regard to that size ball also.

Answering the question as to how the original hardness of the different materials was secured, which was illustrated in the figures given, I might say they were obtained using the pendulum hardness tester, and on the steel balls I think he mentions that he used the steel ball, and on some of the harder material, he used the diamond point. The conversion constant of 10 was used to transfer his reading of the steel ball to the Brinell reading, and 13.5 to transfer the diamond reading to the Brinell reading.

I am not prepared to answer anything in relation to the effect on cast iron and I think it would be only justified in this case that Mr. Herbert be asked to give his views on that particular material.

R. W. WOODWARD: The cloudburst process certainly is an ingenious method for detecting differences in hardness over an entire surface. I would like to inquire a little into the sensitivity of using it in that manner, or rather, the limits to which it can be applied. For instance, it is rather readily seen how it can be used on fairly soft materials where, for instance, we could use the Rockwell "B" scale in inspection work, but I would like to inquire as to whether, in its use on real hard materials, a difference could be detected, say between Rockwell 60 and Rockwell 65 on the "C" scale.

I also wish to emphasize what Captain Petrenko said: that the statement "conversions can be readily made to this scale" should be accepted with limitations, as it has been rather definitely proved by various investigators that the Brinell, Rockwell, Shore, and Pendulum are not inter-convertible among themselves, so it is hardly possible that another method would be convertible to all of these.

In regard to the superhardening process, I would like to inquire if this has been developed in such a way that it can be applied to interior surfaces, such as the wearing surface on a wire-drawing die, or pieces of that nature.

Mr. M. M. give an average that we are that we wondered in examining are quite it is very e

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MR. McMULLAN: Mr. Lewis mentioned that Mr. Herbert's idea was to give an average hardness of a piece. Generally, it is not the average hardness that we are looking for, particularly in a carburized piece, it is the soft spots that we want to be able to pick out rather than the average hardness. I wondered if the machine itself showed those particular soft spots without examining the surface of the piece tested. In carburized pieces, where they are quite often sand-blasted or given a fine shot blast after heat treating, it is very easily seen where the soft spots are located.

R. B. LEWIS: I will take up Dr. Woodward's query. We have some manuscripts from Mr. Herbert and I think I can read Dr. Woodward's answer right there. For instance, if you want to use the cloudburst between limits of 600 to 630 Brinell, 57 to 60 Rockwell, he contends that experience with his cloudburst machine shows that these fine limits have very little meaning. A case hardened disk tested by the cloudburst method presents a map of hardness, commonly exhibiting half a dozen degrees or more of hardness. A Rockwell test made at random on such articles would show it above or below or within limits and cause its acceptance or rejection but is not representative of all its hardness. Some work should not fall below a specified degree of hardness. To meet such cases, the height should be set at which the work is very slightly indented. For example, a cutting tool of 600 Brinell would be slightly indented by three-millimeter balls falling two meters. If it were 650 or 700 Brinell, no indentation would be made. Any soft spots would thus be easily distinguished.

Now, another point about internal work. Cylindrical work can be handled by the cloudburst machine as follows:

The machines are provided with a rotating spindle and with tapered holes in which can be fitted a mandrel for holding gears, bushings, etc. Gears are tested and hardened in gangs and can be set out of center of the shower so the balls fall on the faces of the teeth. Pins can be placed on a roller attachment and rotated. They can also be laid side by side on a 12-inch table and tested on the ends.

The original hardness of these particular gudgeons of which Professor Herbert is speaking was 80 pendulum time-hardness, and after cloudbursting, the pendulum hardness increased to 84, or, in other words, from 800 to 840 Brinell. One of the rings of this lot was apparently of inferior hardness and showed about 750, which was probably produced by excessive grinding.

As to the surface required for cloudbursting, of course it is desirable to remove the scale or any black deposits left by hardening or oil quenching, and the surface left by cylindrical or surface grinding is ideal. The polish originally found on the shank of a drill or tap is equally good and the test shows well on highly polished or lapped surfaces. The chief requirements for rapid and easy inspection are that the surface of the metal should be exposed and the metal should be uniform, not rusty, stained, or otherwise discolored.

I think it would be a little difficult to get inside of a drawing die. As I visualize a drawing die, it has a rather small hole, you could not get into the inside of that very well, but you can get into the inner raceways of ball bearings and material of that shape.

In regard to Mr. McMullan's remarks as to picking out the hard and soft spots, I was unfortunate in the expression used. I should have said it brings into prominence the irregular treatment or unevenness of treatment by indenting the portions which have not received sufficient treatment. Therefore, irregular treatment is brought out, not the average of the hardness, as I stated before.

#### Author's Closure

Replying to the discussion Mr. Herbert wrote that he did not feel it necessary to add much to the remarks of Mr. R. B. Lewis, to whom he was indebted for presenting the paper and replying to the various speakers.

A. L. Davis asked what was the original hardness of the ball bearing, Fig. 4, and of the two balls, Fig. 5. The unworn surface of the ball bearing had a diamond time-hardness of 56.8 equivalent to 768 Brinell. This hardness had been increased by overloading to the equivalent of 885 Brinell. The original surface hardness of the two balls was not known, as they had been superhardened before he received them. The original hardness of steel balls was usually about 850 Brinell by conversion from the pendulum hardness number.

Hard surfaces were not roughened but generally made smoother by the superhardening process. Some ball races which he had received in a highly polished condition were very slightly indented by the commencement of the superhardening process at a height of fall of 2 meters, but when the process had been completed at a height of 4 meters, the surface was free from indentations and more highly polished than before superhardening.

Chilled cast iron was hardened by the pendulum up to a maximum followed by a slight decline as in the case of steel.

Replying to S. N. Petrenko, experiments had been made with balls smaller than 3 millimeters and with various means of propulsion, but it had not been found possible to secure the uniform velocity and the exact regulation of velocity which were essential to a hardness test, otherwise than by gravitation. An adequate velocity could not be imparted to very small balls by gravity without using an excessive height of fall. He was aware that very soft spots could sometimes be detected by sandblasting, but this process was incapable of showing fine gradations of hardness.

He was in general agreement with S. N. Petrenko and also with R. W. Woodward on the subject of conversion to Brinell and other scales of hardness. He was often asked whether the actual hardness could be measured by cloudburst, and he had tested a range of specimens by Brinell and cloudburst and prepared tables of Brinell numbers corresponding to diameters of impressions. Obviously tables could be similarly prepared for conversion to any other scale of hardness, but in practice conversion was rarely resorted to in the case of hard specimens. With a little practice it was possible by inspection to estimate very closely the hardness of a surface which had been tested by the cloudburst. If an absolute measure were required of the hardness at any particular spot, it was, as Mr. Petrenko pointed out, a simple matter to make a Brinell, Vickers, or Rockwell test at that spot.

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# ON THE NATURE OF MARTENSITE CRYSTALS

BY DR. KOTARO HONDA

## Abstract

*The present paper deals with the nature of the martensite, especially with its two kinds, alpha and beta, the existence of which is now made very evident by the result of the X-ray investigation. The formation of these martensites was explained by the theory of quenching put forward some ten years ago. A new view regarding the mechanism of tempering of quenched steels was also expressed. The broadening of the spectral lines of martensites was explained by the distortion of the iron lattice due to the presence of carbon atoms within the lattice.*

## TWO KINDS OF MARTENSITE

**M**ARTENSITE is a structure characteristic to quenched steels and consists of an aggregate of very minute needle-shaped crystals. It is a solid solution of carbon in alpha iron and is very hard. During tempering at a gradually increasing temperature, the precipitation of the carbon in the form of cementite takes place in two steps, that is, at about 265 and 535 degrees Fahr. (130 and 280 degrees Cent.), as seen from the two-stepped changes in the electric resistance-temperature curve, or from a similar anomaly in the magnetization-temperature curve, given in Figs. 1 and 2. Since the precipitation of cementite or the decomposition of martensite takes place at two different temperatures, it is necessary to distinguish two kinds of martensite, alpha and beta, the former being less stable, and hence more easily attacked by acid, than the latter. This interesting fact was first noticed by T. Matsushita<sup>1</sup>; H. Hanemann and Schröder<sup>2</sup> also came to the same conclusion from a microscopic investigation of quenched steels.

By the X-ray analysis, Westgren and others<sup>3</sup> have first found

<sup>1</sup>T. Matsushita, Science Reports of Tohoku Imperial University, Vol. 7, 1918, p. 43.

<sup>2</sup>"On Martensite," by H. Hanemann and A. Schröder, TRANSACTIONS, American Society for Steel Treating, Vol. 9, 1926, p. 169.

<sup>3</sup>Arne Westgren, *Journal, Iron and Steel Institute*, No. 1, 1921; No. 1, 1922; E. C. Bain, *Chemical and Metallurgical Engineering*, Vol. 25, 1921, p. 663.

A paper presented before the tenth annual convention of the society held in Philadelphia, October 8 to 12, 1928. The author, Dr. Kotaro Honda, is honorary member of the society and professor of metallurgy, Tohoku Imperial University, Sendai, Japan. Manuscript received March 27, 1928.

that the martensite has a body-centered cubic lattice, carbon atoms being present within the interspace of the lattice. The position of the carbon atoms in the lattice cannot be determined experi-

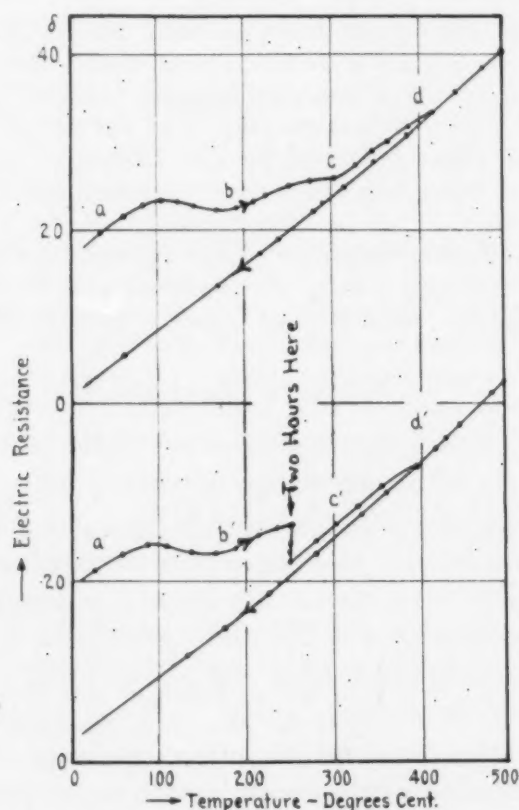


Fig. 1—Curves Showing 2-Stepped Changes in the Electric Resistance-Temperature Relationship (T. Matsushita).

mentally, but is very probably in the center of the face of an elementary cube, distributed here and there according to the law of probability<sup>4</sup>. Recently it was also found that besides the body-centered cubic martensite, a body-centered tetragonal martensite

with an axial ratio  $\frac{c}{a} = 1.03-1.06$  is contained in quenched steel<sup>5</sup>.

S. Sekito and the author<sup>6</sup> found by X-ray analysis that the former

<sup>4</sup>N. Seyjakow, G. Kurdjumow and N. Goodtzow, Science Reports of Tohoku Imperial University, Vol. 14, 1925, p. 170, and Zeitschrift für Physik, Vol. 45, 1927, p. 384.

<sup>5</sup>N. Seyjakow, G. Kurdjumow and N. Goodtzow, loc. cit.

<sup>6</sup>Mr. Weber, Kaiser-Wilhelm Institut, Vol. 6, 1924, Leaflet 1.

<sup>7</sup>Science Reports of Tohoku Imperial University, Vol. 17 (not yet published).

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martensite, which is more stable than the latter and is to be identified with beta martensite, is found in the inner portion of a quenched specimen; while the latter martensite, which is to be identified with alpha martensite, is always found in the surface layer of the specimen. Hence the essential difference between alpha and beta martensites is the distribution of iron atoms in the lat-

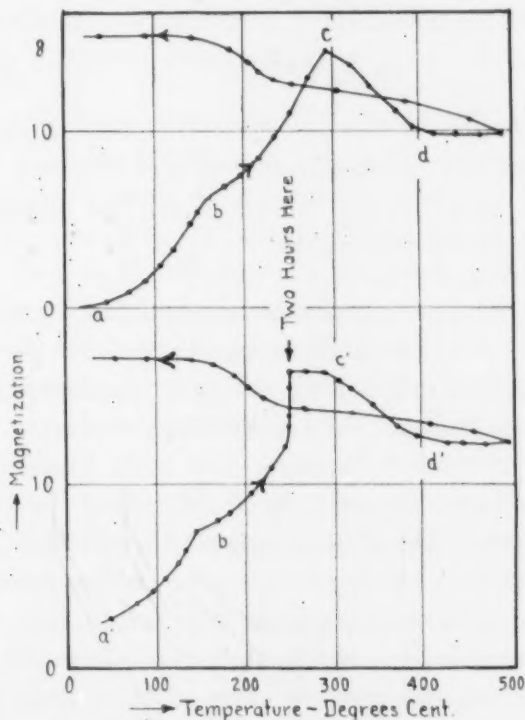


Fig. 2—Curves Showing Magnetization-Temperature Relationship.

tice; in both cases, the carbon atoms being present at the center of the face of an elementary cube or tetragonal, scattered here and there.

Lastly, austenite, which is gamma iron containing carbon in solid solution and stable at high temperatures, has a face-centered cubic lattice, the carbon atoms being present at the center of an elementary cube<sup>7</sup> and distributed according to the law of probability.

#### THEORY OF THE FORMATION OF MARTENSITES

The martensite is obtained by heating steels above the  $A_1$  line

<sup>7</sup>Science Reports of Tohoku Imperial University, Vol. 17 (not yet published).

in the iron-carbon system and rapidly cooling them in water or oil. As to the mechanism of the hardening or of the formation of martensites, there have been a great diversity of opinions, but, thanks to the important results of the X-ray investigation, it is now so clearly understood that there is hardly any doubt as to its formation<sup>6</sup>. This will be briefly shown below:

The  $A_1$  transformation is a change from austenite to pearlite, that is, from the gamma solid solution to a mixture of iron and cementite. Hence the  $A_1$  transformation consists of two changes:

- (1) The change of the atomic configuration of iron from the face-centered to the body-centered lattice;
- (2) The separation of carbon atoms as cementite from the interspace of the lattice.

These two processes take place very probably in rapid succession, but not exactly at the same moment; because in the case of a natural phenomenon, it is very difficult to consider two independent processes as taking place exactly at the same moment. Admitting these successive changes, the next question is which of the changes (1)  $\rightarrow$  (2), or (2)  $\rightarrow$  (1), is more probable? The answer is obviously the first change (1)  $\rightarrow$  (2); because, if the change in the atomic configuration of iron from the face-centered to the body-centered is completed, the product is the alpha iron dissolving carbon, but the alpha iron not being able to dissolve so much carbon in it, the greater part of the latter is immediately precipitated as cementite. Hence if the first change occurs, the second one must follow as a necessary consequence, that is, the order of the change (1)  $\rightarrow$  (2) is very probable. If this view be correct, the process of the  $A_1$  transformation, that is,

austenite  $\rightarrow$  pearlite

is that

gamma iron dissolving carbon  $\rightarrow$  alpha iron dissolving carbon  $\rightarrow$  ferrite +  $Fe_3C$ ,

or

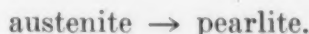
austenite  $\rightarrow$  martensite  $\rightarrow$  pearlite.

That is, the  $A_1$  transformation consists, in process, of double

<sup>6</sup>Dr. Kotaro Honda, Science Reports of Tohoku Imperial University, Vol. 14, 1925, p. 165.

or stepped changes, the martensite being an intermediate stage of the transformation from austenite to pearlite.

In the case of slow cooling at the  $A_1$  point, austenite first changes into martensite, and the latter changes immediately into pearlite, the result being the same as



In the case of a very rapid cooling, such as quenching in water, the change from austenite to martensite is so far retarded that it begins to take place at a temperature below 570 degrees Fahr. (300 degrees Cent.) and when this change is completed, the specimen which is subjected to this treatment, is nearly at room temperature. Hence, the second change from martensite to pearlite cannot take place; because this change involves the diffusion of carbon atoms through iron and therefore the specimen will undergo a high resistance owing to its greater viscosity at room temperature. Thus the martensite is obtained by quenching the steel in water.

According to the above theory, the perfect quenching or hardening is obtained when the first half of the  $A_1$  transformation, that is, austenite  $\rightarrow$  martensite is completed, and the second half, martensite  $\rightarrow$  pearlite, is completely suppressed. In the case of a less rapid cooling, the first half of the transformation occurs at a temperature a little higher than in the above case, and therefore the second half partly takes place, resulting in a martensitic structure mixed with troostite. Troostite is a dispersed system, in which cementite is suspended in iron as very fine colloid-like particles, and hence is a mechanical mixture of cementite and iron, and does not differ physically from pearlite. Since the hardness,  $H$ , of austenite, martensite and troostite are in the following order

$$H_{\text{martensite}} > H_{\text{troostite}} > H_{\text{austenite}},$$

the hardness of martensite mixed with troostite is less than that of pure martensite. In this case there is an imperfect hardening.

In the case of an extremely rapid cooling, not only is the second half of the  $A_1$  transformation completely suppressed, but also the first half is partially arrested. We then have a martensite

mixed with austenite, the hardness of which is less than that of pure martensite. This is the case of a too severe quenching. That in a severely quenched steel, a small amount of austenite is present mixed with martensite, is a fact actually confirmed by X-ray analysis.

The nature or mechanism of the quenching being so far made clear, we shall show below in what relation the quenched structure having a body-centered tetragonal lattice stands to that having a body-centered cubic.

Since a face-centered cubic lattice may be considered a body-centered tetragonal one with an axial ratio  $\frac{c}{a} = \sqrt{2}$  and also a body-centered cubic lattice as a body-centered tetragonal lattice with an axial ratio  $\frac{c}{a} = 1$ , the transformation from austenite to martensite takes place very probably in the order:

$$\text{Tetragonal lattice } \left( \frac{c}{a} = \sqrt{2} \right) \rightarrow \text{tetragonal lattice } \left( \frac{c}{a} = 1.06 \right) \rightarrow \text{tetragonal lattice } \left( \frac{c}{a} = 1 \right),$$

or, austenite  $\rightarrow$  alpha martensite  $\rightarrow$  beta martensite.

Thus the mechanism of the formation of these martensites from austenite is very simple; that is, the tetragonal lattice  $\left( \frac{c}{a} = \sqrt{2} \right)$  is first to be compressed uniformly in the direction of the c-axis and at the same time uniformly expanded in the perpendicular direction. Alpha martensite is obtained when the axial ratio of the tetragonal changes from  $\sqrt{2}$  to 1.06, and beta martensite when the ratio reduces farther to 1. During these changes, it is not necessary for the carbon atoms in the lattice to change their position relative to the iron atoms; because, as seen from Fig. 3, the carbon atoms in austenite occupy the center of the elementary cube, and those in martensite that of the face of the cube, so that the carbon atoms remain in their position during these changes<sup>9</sup>.

<sup>9</sup>N. Seyjakow, G. Kurdjumow and N. Goodtzow, *Zeitschrift für Physik*, loc. cit.



In the outer layer of a quenched steel, where the cooling is very rapid, the first change, austenite  $\rightarrow$  alpha martensite, is partially arrested and the second change, alpha martensite  $\rightarrow$  beta martensite, is completely hindered, so that the outer layer must contain alpha martensite mixed with a small quantity of retained austenite; while in the inner portion, where the cooling is less rapid, the first and the second change will take place almost com-

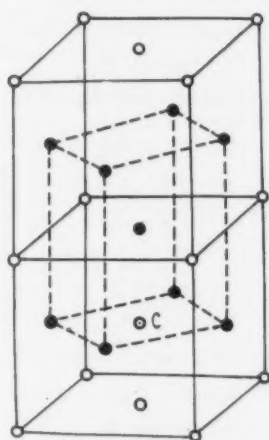


Fig. 3—Diagram Showing Manner in Which Carbon Atoms in Austenite Occupy the Center of the Elementary Cube and Thus the Martensite That of the Base of the Cube so That the Carbon Atoms Remain in Their Position During These Changes.

pletely, so that the inner portion must contain almost pure beta martensite. These conclusions agree satisfactorily with the results of the experiment made by S. Sekito and the author.

#### PROPERTIES OF ALPHA AND BETA MARTENSITES

##### (a) *Lattice Constants*

S. Sekito and the author took the X-ray spectra of quenched steels containing different percentages of carbon, and obtained the lattice constants of austenite, alpha and beta martensites, as shown in Table I:

##### (b) *Specific Volumes*

The specific volume  $V$  of these structures can be found from the lattice constants above given by the following formula:

$$v = \frac{V}{1.649 \times 10^{-24} W n},$$

where  $V$  is an elementary volume constituting the space-lattice,  $1.649 \times 10^{-24}$  is the mass of one hydrogen atom divided by 1.008,  $W$  the atomic weight, and  $n$  the number of atoms in the elementary volume. Table II contains the specific volumes for austenite, and for the alpha and beta martensites:

Table I.  
Lattice Constants

Carbon content	Austenite $a$	Alpha martensite $a$ $c$		Beta martensite $a$
0.20	3.596	2.839	1.068	2.860
0.40	3.610	2.838	1.073	2.863
0.51	3.604	2.836	1.072	2.864
0.62	3.604	2.832	1.073	2.864
0.73	3.604	2.834	1.072	2.867
0.89	3.607	2.837	1.066	2.870
1.04	3.610	2.837	1.066	2.866
1.08	3.607	2.839	1.055	.....

Table II.  
Specific Volumes

Carbon content	Austenite	Alpha martensite (tetragonal)	Beta martensite (cubic)
0.20	0.1257	0.1327	0.1271
0.40	0.1260	0.1332	0.1275
0.51	0.1259	0.1327	0.1276
0.60	0.1259	0.1324	0.1276
0.73	0.1259	0.1324	0.1280
0.89	0.1259	0.1321	0.1284
1.04	0.1260	0.1323	0.1279
1.07	0.1259	0.1313	.....

In Fig. 4, the specific volumes of these constituents are plotted against the carbon content. The specific volume of austenite and the beta martensite slightly increases, but that of the alpha martensite decreases a little, with the increase of carbon content. These changes are given by the following formula:

$$\begin{aligned} \text{austenite: } V &= 0.1257 + 0.003 C, \\ \text{alpha martensite: } V &= 0.1335 - 0.0018 C, \\ \text{beta martensite: } V &= 0.1267 + 0.0017 C, \end{aligned}$$

$C$  being the weight per cent of carbon. The rate of expansion of austenite due to carbon is less than that found by Wever<sup>10</sup> in the

<sup>10</sup>Weaver, *Mitteilungen aus dem Kaiser-Wilhelm Institut*, Vol. 6, 1924, Leaflet 1.

case of manganese iron. Thus the specific volumes of austenite, alpha and beta martensites satisfy the following relation:

$$V_{\text{austenite}} < V_{\beta \text{ martensite}} < V_{\alpha \text{ martensite}}$$

According to our theory of quenching, in a severely quenched steel, the inner portion may contain austenite, alpha and beta mar-

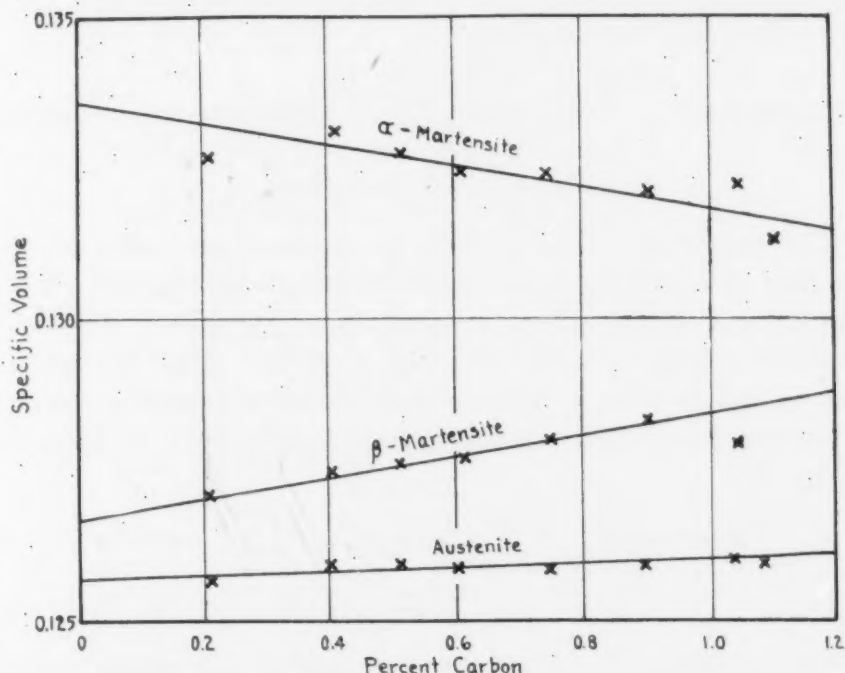


Fig. 4—Curves Showing the Specific Volumes of Austenite and of Alpha and Beta Martensite.

tensites; but actually we find austenite and beta martensite, and not alpha martensite. This absence of alpha martensite is caused by the large specific volume of the constituent; because in the inner portion of the specimen, the formation of alpha martensite is resisted by a great pressure exerted from the outer portion of it, and hence either the change of austenite into alpha martensite does not take place, or if it does, is passed over to beta martensite. By a recent investigation it is also confirmed that in the outermost layer of a quenched steel, the alpha martensite has an axial ratio of about 1.07-1.06, and as the depth from the surface increases, this ratio rapidly decreases and soon attains its asymptotic value 1.00.

(c) *Magnetizability*

Austenite is paramagnetic, and martensite ferromagnetic, though it is weaker than in the case of pearlite. Since the space lattice of alpha martensite is an intermediate one between austenite and beta martensite, and is very near to that of the latter structure, it is to be concluded that alpha martensite is also ferromagnetic, but less magnetic than beta martensite. Hence the magnetizability  $I$  of austenite, alpha and beta martensites and pearlite satisfy the following relation:

$$I_{\text{austenite}} < I_{\alpha \text{ martensite}} < I_{\beta \text{ martensite}} < I_{\text{pearlite}}$$

(d) *Electric Resistance*

For carbon steels, the specific resistance of austenite is greater than that of beta martensite; for example in the case of a 0.5 per cent carbon steel, the specific resistances of austenite and beta martensite are respectively  $32.7$  and  $20.8 \times 10^{-6} \Omega$ . Since the structure of alpha martensite is intermediate between austenite and beta martensite, the electric resistance must also lie between these two; that is,

$$R_{\beta \text{ martensite}} < R_{\alpha \text{ martensite}} < R_{\text{austenite}}$$

(e) *Hardness*

The hardness of a substance depends, among several factors, on the atomic distance or the specific volume; thus when other conditions are equal, the hardness is greater as the specific volume is smaller. The only important difference between alpha and beta martensites is their specific volume; hence the hardness of these structures must satisfy the following relation:

$$H_{\text{austenite}} < H_{\alpha \text{ martensite}} < H_{\beta \text{ martensite}}$$

Now martensites have nearly or exactly the same atomic configuration as that of ferrite, the only difference being the presence of carbon atoms in the interspace of the lattice; notwithstanding they are very much harder than ferrite. The cause of this abnormality is due to several factors, that is, the fineness of grain size,

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internal stress and the presence of carbon atoms within the lattice. It is a well established fact that the hardness of a metal increases with the grain number, and that cold work resulting to internal stress causes also the hardening of the metal. The martensites consist of an immense number of extremely fine crystals, and also are undergoing a great internal stress caused by a rapid cooling during quenching. Again, carbon atoms in the lattice will send their lines of force toward six neighboring atoms and themselves act like diagonal supporters in a square framework (Fig. 5),

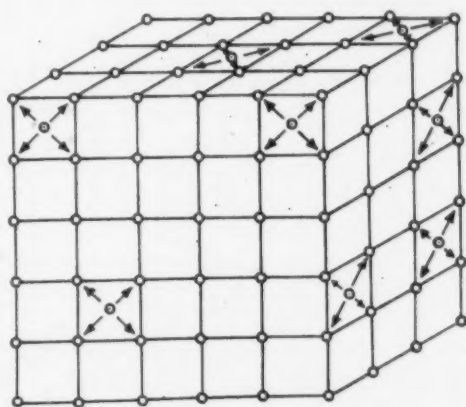


Fig. 5—Diagram Showing the Manner in Which the Carbon Atoms in the Space Lattice Will Send Their Lines of Force Toward 6 Neighboring Atoms and Themselves Act Like Diagonal Supporters in a Square Framework, Thus Giving Great Strength to the Lattice as a Whole.

thus giving to the lattice building a great strength as a whole. Since the effect of grain size and that of internal stress on the hardness are fairly well known, it is possible to estimate approximately quantitative contributions of these factors to the hardness, that is, 80 in Brinell number for the grain size and 150 for the internal stress are probably not far from the truth. Assuming the Brinell hardness for a 0.9 per cent carbon steel quenched to be 680 and that for the same steel annealed to be 220, we get

Natural hardness of the steel .....	220
Increase due to grain number .....	80
Increase due to interstrain .....	150
Increase due to carbon atom within lattice .....	230
	<hr/>
	680

Many metallurgists have hitherto sought for the cause of the

great hardness of martensite, and concluded it to be chiefly due to either one of these factors; but this view cannot be correct.

#### MECHANISM OF THE TEMPERING OF MARTENSITES

It is usually admitted that the tempering of quenched steels takes place at two steps, as shown in Figs. 1 and 2 in the first section. The same fact may also be seen from Figs. 6 and 7, which show the thermal analysis and the thermal expansion of quenched

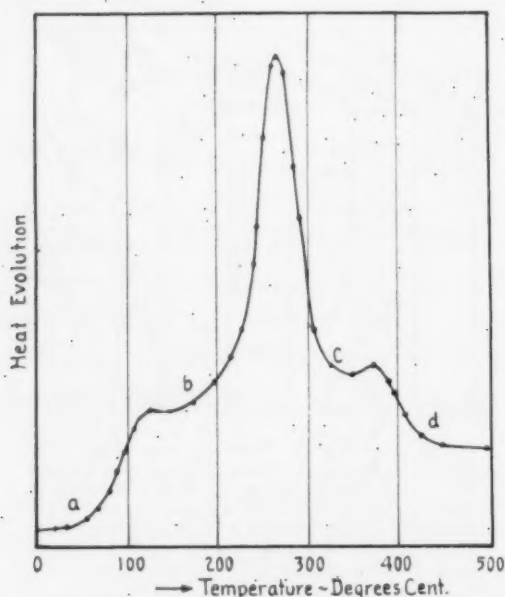


Fig. 6—Curves Showing Thermal Analysis of Quenched Steels.

steels. One year ago, even the present writer considered it as an established fact; but recent experiments in different directions have led him to the following different view.

The first step at about 265 degrees Fahr. (130 degrees Cent.) in the curves of magnetization, electric resistance, heat evolution and thermal expansion plotted against the tempering temperature is not caused by the decomposition of martensite or the precipitation of cementite, but by the change of alpha into beta martensite. The decomposition of beta martensite is assumed to begin with below 212 degrees Fahr. (100 degrees Cent.) its rate being at first very small and rapidly increasing with the rise of temperature, and is completed at about 535 degrees Fahr. (280 degrees

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Cent.), the second step in Figs. 1, 2 and 6 being due to this rapid decomposition of the austenite. In other words, the first step in these figures is not the abrupt increase in the rate of decomposition of the martensite, but shows the change of alpha into beta martensite.

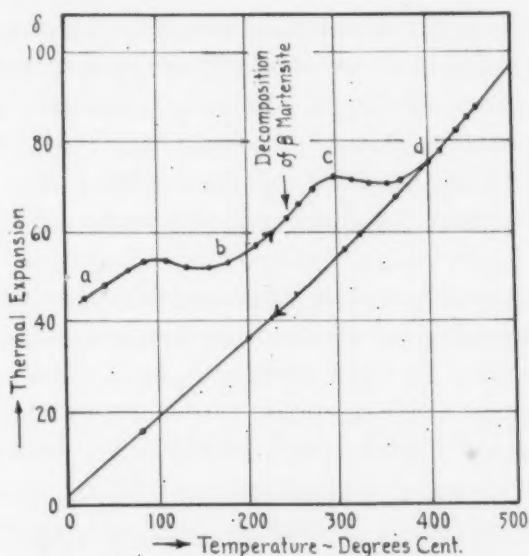


Fig. 7—Curves Showing Thermal-Expansion of Quenched Steels.

As we have seen,

$$\begin{array}{lcl}
 R_{\alpha \text{ martensite}} & > & R_{\beta \text{ martensite}}; \quad I_{\alpha \text{ martensite}} < I_{\beta \text{ martensite}}; \\
 \text{and} & & \\
 V_{\alpha \text{ martensite}} & > & V_{\beta \text{ martensite}}
 \end{array}$$

and hence it is to be expected that during the change alpha to beta martensite, the electric resistance and the length of the specimen diminish, while the magnetization increases. Since during the same change, the volume of the specimen contracts, the heat must be evolved. These conclusions are borne out by actual experiments as seen from Figs. 1, 2 and 6.

The fact that by tempering a quenched steel at gradually increasing temperature, the hardness increases up to about 250 degrees Fahr. (120 degrees Cent.), is also attributed to the change of alpha into beta martensite. Since beta martensite is harder than alpha martensite, the change of alpha to beta martensite re-

sults in the increase of hardness; but from the same range of temperature upwards beta martensite is more or less decomposed, this resulting in the decrease of hardness. As the result of these two opposite effects, the hardness first increases, attains a maximum and then decreases.

The second step be in the curves of Figs. 1, 2 and 6 is obviously due to the rapid decomposition of beta martensite, thereby the electric resistance and the magnetization increase and the heat of precipitation is evolved. But curiously enough, in the same interval of temperatures, 390 to 535 degrees Fahr. (200 to 280 degrees Cent.) in which the above change takes place, the change in length is not observable; the cause of this absence will be explained afterwards.

During tempering, carbon atoms within the lattice will move toward grain boundaries, where they are set free. The present writer<sup>11</sup> assumes that all these carbon atoms do not at once take up iron atoms to form the cementite, but some of them remain uncombined for some interval of temperature. Even the newly formed cementite cannot immediately build up the lattice characteristic to the cementite, so that its property is considerably different from that of ordinary cementite and probably near that of iron. The main building up of cementite will take place in a temperature interval of 570 to 750 degrees Fahr. (300 to 400 degrees Cent.). The third step ed in the curves of Figs. 1, 2 and 6 corresponds to this building up of the ordinary cementite lattice from the colloid-like particles of cementite. During this change the magnetization must decrease, since above 420 degrees Fahr. (215 degrees Cent.) cementite is already nonmagnetic. The fact that the magnetization-tempering temperature curve shows hardly the A<sub>0</sub> transformation by tempering below 570 degrees Fahr. (300 degrees Cent.) is also attributed to the incompleteness of the cementite lattice, which is chiefly formed at 570 to 750 degrees Fahr. (300 to 400 degrees Cent.). During the same change, as the atoms take gradually a regular arrangement, the electric resistance as well as its specific volume must decrease. Since the change of atomic configuration from an irregular to a regular stable one involves the diminution of the potential energy, this must appear

<sup>11</sup>"On the Mechanism of Tempering of Steels," by Tokujiro Matsushita and Kiyoshi Nagasawa, Science Reports of Tohoku Imperial University, Vol. 16, 1927, p. 901. Writer's view is similar to that of Mr. Matsushita and Mr. Nagasawa, but different in some respects.

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as the heat energy. These conclusions agree quite satisfactorily with the observed facts.

In the case of a 1 per cent carbon steel, the specific volumes of martensite, the mixture of ferrite and cementite and that of ferrite and carbon are respectively 0.1283, 0.1276 and 0.1303. Hence the decomposition of beta martensite into ferrite and cementite should accompany a decrease of volume, and that of the martensite into ferrite and carbon an increase of volume; but since during the decomposition, no change of volume is observable, the great part of the decomposition products must be ferrite and cementite. At the moment of the decomposition of beta martensite, the cementite particles do not take their own specific volume, but are in a loosely packed state, leaving minute gaps between their contact surfaces, so that the whole volume remains apparently undiminished during the decomposition of the martensite, as actually observed.

#### BROADENING OF THE X-RAY SPECTRAL LINES

One important point to be noticed is the broadening or diffusion of the X-ray spectral lines of martensite. This is usually explained by assuming that each individual needle-shaped crystal consists of a large number of minute crystals orientated at random, their linear dimensions containing only several hundreds of atoms. That explanation does not, however, agree well with the result of microscopic observation, but the following new exposition which was first given by the writer three years ago, would very probably express the actual fact.

It is a well known fact that alpha iron at room temperature cannot contain any carbon in solid solution, and hence the carbon atoms in martensite are present in a strongly forced condition, their precipitation being hindered by a great viscosity of the metal at room temperature. Hence in the neighborhood of the portions, where carbon atoms occupy their positions in the interspace of the iron lattice, the latter is highly strained, the degree of straining rapidly decreasing as the distance from the carbon atoms increases. The result is that different portions of the iron lattice are undergoing different degrees of internal strain or stress, the degree of straining or stress in the lattice continuously decreasing as the distance from the carbon atoms in the lattice increases. This con-

tinuous variation of the lattice parameter with carbon atoms as centers will cause the broadening of the spectral lines. In the case of a quenched steel, containing 1.0 per cent carbon, the mean expansion of the lattice was 0.45 per cent and the broadening of the lines 1.0 per cent of the lattice parameter<sup>12</sup> it is therefore to be concluded that the presence of carbon atoms, 1.0 per cent in amount, in the interspace of the lattice resulted in the deformation of the iron lattice into an immense number of minute tetragonal lattices, whose axial ratio varies continuously from 1.00 to 1.01. Thus the diffusion of the spectral lines is satisfactorily explained by the deformation of the lattice by the presence of carbon atoms within it, there being no need to assume a very finely dispersed structure in each of the martensite crystals.

That the broadening of the spectral lines is not attributed to a dispersed structure of martensite, is also confirmed by the result of experiments on the effect of annealing quenched steels on the breadth of the spectral lines, that is, by the fact that by annealing at 570 degrees Fahr. (300 degrees Cent.) for 30 minutes, the broadening of the lines nearly vanishes. Since the grain or crystal growth cannot take place, at such a low temperature, it is evident that the broadening of the lines is not due to the fineness or the size of the crystal.

Again in the spectral lines belonging to martensite, several dark spots are observable as in the case of a cold-worked iron, showing that the martensite must also contain crystals of a considerable size.

It is interesting to calculate the internal stress existing in quenched specimens from the broadening of the spectral lines above referred to. Assuming Young's modulus of elasticity of quenched steel (1.0 per cent carbon) to be  $2 \times 10^4$  kilograms per square millimeter, the maximum stress corresponding to the above broadening of lines becomes

$$P = 2 \times 10^4 \times 0.01 = 200 \text{ kilograms per square millimeter}$$

which agrees with the value of tensile strength of the quenched steel. Since internal stress cannot increase beyond the breaking strength of a material, the above result is quite plausible.

<sup>12</sup>Dr. Kotaro Honda and S. Sekito, loc. cit.

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## DISCUSSION

**Written Discussion:** By Dr. Albert Sauveur, Harvard University, Cambridge, Mass.

Professor Honda's opening sentence to the effect that martensite is a structure characteristic of quenched steel and that it consists of an "aggregate" was good reading as I have for some time past repeatedly expressed a similar view.<sup>1</sup> I shall welcome the day when martensite will cease to be described as a constituent, generally as a solid solution of carbon in alpha iron. I was, therefore, disappointed when in his third line the author also describes martensite as a solid solution of carbon in alpha iron. It is obvious that Professor Honda does not use the word "aggregate" as most of us are in the habit of doing, namely, as signifying the mixture of at least two phases. He states further that in tempering, martensite is converted into an aggregate—a real one this time—through the precipitation of cementite, which must mean, I believe, the formation of troostite.

We must agree with Professor Honda when he states that the decomposition of austenite involves two steps: (1) the allotropic transformation of gamma into alpha iron, and (2) the precipitation of minute particles of cementite. Indeed, has not this view been held by many for a considerable length of time? We also readily accept his opinion that in all probability the allotropic transformation precedes the precipitation of cementite. It also seems evident that the two transformations are not instantaneous. The first one implies the formation of an alpha phase at the boundaries and along the crystallographic planes of the austenite grains. This alpha phase may originally consist of a solid solution of carbon in alpha iron, but because of the slight solubility of carbon in alpha iron the second transformation sets in, namely, the rejection by the alpha phase of minute carbide particles by which it is converted into troostite. By arresting the transformation at this stage, as is done in quenching, the steel is preserved in the form of an aggregate of austenite and troostite. This, I believe, is the nature of the structure generally called martensite. The acicular character of this structure results from the alpha phase locating itself along the crystallographic planes of the original austenite, as is so well illustrated by Lucas.<sup>2</sup>

On slow cooling, the alpha phase, assuming the steel to be hypoeutectoid, is completely freed from carbon which is absorbed by the surrounding gamma phase because of the ability of gamma iron to dissolve as much as 0.90 per cent carbon below the  $A_{cm}$  point. The alpha phase is thereby converted into what is known as free ferrite. The gamma phase or austenite on reaching the  $A_1$  point becomes of eutectoid composition and on further cooling is transformed into pearlite. The resulting structure will present the well-known Widmanstätten pattern which, in my opinion,

<sup>1</sup>Albert Sauveur, "The Widmanstätten Structure of Iron Alloys," *Proceedings, American Philosophical Society*, Vol. LXVI, 1927.

<sup>2</sup>Francis F. Lucas, "The Microstructure of Austenite and Martensite," *TRANSACTIONS, American Society for Steel Treating*, Vol. VI, 1924, pp. 18 and 19.

is of the same type as the martensitic pattern and results from similar causes. It will suffice to remember that a martensitic structure is formed quickly at a low temperature (some 300 degrees Cent.), while the Widmanstätten structure forms leisurely at some 900 to 700 degrees Cent., to understand why the alpha phase in the former, that is the troostite needles, should be so much finer than the corresponding constituent, namely, the free ferrite in the latter. As to the hardness of the martensitic structure, it may be due to its fineness, to the retained austenite being an under-cooled solid solution, to some carbon possibly remaining dissolved in the alpha iron of troostite in the form of a supersaturated solid solution and to severe internal strains.

Some of us, I believe, will find it difficult to accept Professor Honda's evidence as conclusive of the existence of two forms of martensite, namely, alpha and beta martensite. While some investigators have reported the existence of two types of space lattices, the body-centered cubic and the body-centered tetragonal, others, also skilled and experienced, have been unable to detect the occurrence of the tetragonal lattice. E. C. Bain believes that the tetragonal lattice may be merely an arrangement temporarily assumed by a face-centered lattice in transforming into a body-centered one.

**Written Discussion:** By Dr. R. L. Dowdell, Bureau of Standards, Washington.

Dr. Honda is to be congratulated on his many contributions to the physical metallurgy of steel. His present paper on the nature of martensite crystals will be received with a great deal of interest by those who have studied this very complicated problem.

After carefully studying Dr. Honda's paper, there are some points which appear to be open for further discussion, particularly when considered with the results of other researches. In the first place many investigators of steels can see no reason why both martensite needles and troostite needles should be redesignated and called alpha and beta martensites, which apparently Dr. Honda has done. When Professor Osmond named the white acicular structure occurring in quenched steels martensite it was understood that this white acicular needle-like material had different physical and chemical properties from the darker etching material which he named troostite. Quoting from Stead's translation<sup>3</sup> of Osmond's book on "The Microscopic Examination of Metals" is it stated: "Martensite, isolated by polish attack, is generally colorless, or only a pale yellow. On the other hand, the first drop of tincture of iodine causes a yellow brown, or black coloration, according to the amount of carbon, the fibers always coming out bright."

Most metallurgists have used these terms ever since to designate these microconstituents. In other words, the so-called beta martensite is the same as troostite because on formation from austenite during quenching or on tempering, it apparently has the same physical properties and etching characteristics.

Dr. Honda still retains the idea that the more severe the quenching of

<sup>3</sup>Second Edition, published by Charles Griffin and Co., Ltd.

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<sup>7</sup>Harde  
1927, p. 5



steel the greater the amount of austenite, but this idea seems contrary to fact because the more severe the quenching the greater the stress and the greater will be the amount of martensite with the greater degree of hardness. This relation has been pointed out by Dr. John A. Mathews; and by the writer with O. E. Harder.<sup>5</sup>

The tempering of quenched steels by heating for long periods of time at a relatively low temperature, 100 degrees Cent. (212 degrees Fahr.), will convert the (white needle-like) martensite into the darker troostite which has all of the same characteristics as the troostite (beta martensite) formed at about 270 degrees Cent. (535 degrees Fahr.).

In other words, if the conditions of heating during tempering approach stable equilibrium, i. e., are extended over long periods of time, we have found that the lattice parameters ( $a$ ) of the so-called alpha and beta martensites or any mixture of these products on low and high temperature tempering will vary between 2.839 and 2.861 with no apparent order, dependent on either temperature or carbon content. This conclusion is based upon some results which have been published<sup>6</sup> and others as yet unpublished.

Under tempering conditions approaching stable equilibrium these martensite needles gradually grade into the troostitic condition so that it does not seem proper to give lattice constants for these variations<sup>7</sup> as shown in Table I.

Dr. Honda still assumes that on slow cooling of eutectoid steels from above the  $A_1$  point the decomposition of austenite to pearlite takes place through the step martensite. From experimental evidence it will be found, as Dr. Honda has stated, that martensite will be formed at a temperature of about 300 degrees Cent. (570 degrees Fahr.) or below if the quenching is fast enough. If the quenching is somewhat slower some of this martensite is tempered to troostite which in regard to the phases present is analogous to pearlite.

If the cooling is slow, such as air cooling, it does not seem that pearlite can form direct from martensite because martensite will not be formed at the high temperature of 700 degrees Cent. where the steel is plastic. In order to form martensite we must have stress and mainly tensile stress. When austenite decomposes on cooling it appears reasonable to say, as Dr. Honda and others<sup>7</sup> have, that it takes place in the successive stages: (1) the change of atomic configuration of iron from the face-centered to the body-centered lattice, and (2) the separation of carbon atoms as cementite from the interspaces of the lattice.

It seems that on air cooling it will be impossible for the austenite

<sup>4</sup>Mathews, J. A., "Retained Austenite—A Contribution to the Metallurgy of Magnetism," TRANSACTIONS, American Society for Steel Treating, Vol. VIII, 1925, pp. 565-583.

<sup>5</sup>Dowdell, R. L., and Harder, O. E., "The Decomposition of the Austenitic Structure in Steels, Part I," TRANSACTIONS, American Society for Steel Treating, Vol. XI, 1927, p. 217.

<sup>6</sup>Harder and Dowdell, TRANSACTIONS, American Society for Steel Treating, Vol. XI, 1927, p. 959.

<sup>7</sup>Harder and Dowdell, TRANSACTIONS, American Society for Steel Treating, Vol. XII, 1927, p. 51.

to go through the step martensite because in the strict sense martensite, as used by previous investigators, must be formed at lower temperatures where the stress is high and a metastable condition exists. Knowing that the greatest amount of martensite is formed on the highest speeds of cooling and that on extremely slow tempering austenite transforms to nodular troostite without even the formation of needles (unless there is a fairly high temperature gradient in the piece thus causing stress), it seems that on slow cooling austenite should not go through the martensite step.

**Written Discussion:** By Dr. W. L. Fink, Aluminum Company of America, Cleveland.

In his theory of the nature of martensite, Dr. Honda assigns an important rôle to the body-centered tetragonal lattice. It is held responsible for all changes brought about by the aging of quenched steel, and is supposed to account completely for the broadened lines in the X-ray patterns of martensite. The question naturally arises as to how much we know about this body-centered tetragonal lattice.

The data presented in this article confirm the results of two previous investigations<sup>8,9</sup> in that there exists in some hardened steels a body-centered tetragonal structure with an axial ratio of approximately 1.06-1.00, which changes over to a body-centered cubic structure at a temperature of 265 degrees Fahr. (130 degrees Cent.) or less. The specific volume of this structure is calculated from the lattice parameters, and the parameters found in all three investigations place the specific volume at 0.133 or less.

There is also agreement as to the specific volumes of the body-centered cubic structure (approximately 0.127) and the face-centered cubic structure (approximately 0.126).

There is a disagreement as to the effect of carbon content on the parameters *a* and *c*. Honda's results indicate that the axial ratio decreases as the carbon content increases, whereas Seljakow, Kurdjumow, and Goodtzow found that the axial ratio increased as the carbon content increased.

Practically nothing is known about the other properties of the tetragonal structure. Honda assumes that they will be intermediate between those of austenite and "beta martensite". Such reasoning would lead to an erroneous result if applied to the specific volume, and may be inaccurate when applied to electric and magnetic properties.

On page 105 of the paper the following statement is found:

"By a recent investigation it is also confirmed that in the outermost layer of a quenched steel, the alpha martensite has an axial ratio of about 1.07-1.06, and as the depth from the surface increases, this ratio rapidly decreases and soon attains its asymptotic value 1.00."

<sup>8</sup>Wm. L. Fink, "Influence of Heat Treatment and Carbon Content on the Structure of Pure Iron-Carbon Alloys," TRANSACTIONS, American Society for Steel Treating, Vol. IX, 1926, p. 717.

<sup>9</sup>Kurdjumow, Seljakow, and Goodtzow, "Eine Röntgenographische Untersuchung der Strukturen des Kohlenstoffstahls," *Zeitschrift für Physik*, Vol. 45, 1927, p. 384.

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<sup>10</sup>W. F.  
TRANSACTIONS

If this statement and Honda's theory were both true, it would be possible to grind off the outer layers of a piece of quenched steel and thus eliminate all aging effects. This offers an experimental method of verifying Honda's theory.

There is at least one feature of the aging of steel for which Honda's theory does not account. Sykes and Jeffries have shown<sup>10</sup> that upon aging hardened steels at 50 degrees Cent. the hardness increases to a maximum at about 150 hours and then decreases. At higher temperatures this maximum is obtained in a shorter time. The shape of these aging curves, and the effect of temperature on the aging velocity are similar to those of the strong aluminum alloys, and can be explained on the basis of the precipitation theory of hardening. Honda's theory does not account for the softening which is obtained after maximum hardness is reached.

#### Author's Reply

My hearty thanks are due to Dr. Sauveur, Dr. Dowdell and Dr. Fink who have been so kind to take part in the written discussion of my paper. The main points of divergence between the opinion of these gentlemen and ours are to be attributed to the difference of the definition of martensite and troostite, which exists between us. Years ago, when the microscope was the only means for studying the structure of metals and alloys, the nature of martensite and troostite was very obscure; but thanks to the important results of the X-ray investigation, it now becomes so clear that we can discuss these structures from the viewpoint of atomic configuration, and hence the old definitions regarding these structures should be, in my opinion, properly modified. Accordingly throughout my paper, the following definitions have been strictly followed:

(1) Martensite is an iron having a body-centered cubic lattice or a body-centered tetragonal lattice having an axial ratio of about 1.06, with carbon atoms within the lattice.

To distinguish between the above two kinds of martensites, the former is designated as beta martensite and the latter alpha martensite. Under the microscope, martensite appears to consist of acicular needle-shaped crystals, of which the one, that is slightly colored by etching with acid, is alpha martensite and the other that is colored darker, the beta martensite. The needle-shaped appearance of martensite crystals is not to be taken as an essential criterion for martensite, as the appearance itself may be greatly modified with varying conditions of formation.

(2) Troostite is a mechanical mixture of ferrite and cementite, these cementite particles being so finely dispersed as to form colloid-like particles.

Under the microscope, troostite is, as etched with acid, a dark-colored structure, of which the individual crystals cannot be distinguishable.

According to the above definitions, beta martensite in my case is a troostite in the sense of Sauveur, Dowdell and Osmond. Dr. Sauveur seems to define "martensite" as a mixture of austenite and troostite

<sup>10</sup>W. P. Sykes and Zay Jeffries, "On the Constitution and Properties of Hardened Steel," TRANSACTIONS, American Society for Steel Treating, Vol. XII, 1927, p. 871.

(beta martensite in my definition); but we cannot agree with this nomenclature. I believe, it is better to preserve the name "martensite" for the homogeneous alpha or beta solid solution, as we have defined above.

As to the causes of hardness of martensite, Dr. Sauveur enumerates one more, that is, retained austenite besides the three which I have mentioned; but in my definition of martensite, the retained austenite is excluded, and therefore it is not necessary to take it into account. Since, however, austenite is much softer than pearlite, the retained austenite in martensite greatly reduces the hardness of the mixture.

It is nowadays a fact beyond doubt that in a quenched steel, two kinds of martensite lattices, alpha and beta, are found; the former exists only in the outer layer of the steel, and the latter in the inner portion. Westgren could find only beta martensite; because in his experiment the surface layer of the specimen was filed off. E. C. Bain's view regarding the alpha martensite is exactly the same as mine, alpha martensite being a transition stage from austenite to beta martensite.

For the greater part of the objections raised by Dr. Dowdell, the above explanation, I believe, gives my answer; a few points remain to be explained.

In the case of carbon steel, it is a matter beyond doubt that as the quenching is severe, a greater amount of austenite is found mixed with martensite. This is repeatedly confirmed by X-ray analysis. In a recent investigation in my Institute, which is not yet published, the different portions of a quenched specimen were subjected to the X-ray analysis and the intensity of spectral lines belonging to martensite and austenite measured by means of a microphotometer, the relative amount of these structures being thus determined. The same process was applied to both oil and water-quenched specimens of steels containing different amounts of carbon. The result showed that the amount of austenite is greater in the outer portion than in the inner, and that the martensite is present in a greater amount in the water-quenched specimen than in the oil-quenched.

According to our result of the X-ray investigation,<sup>11</sup> during tempering of a quenched steel, the lattice constant of alpha martensite does not change continuously into that of beta martensite but the change is quite abrupt. Hence it is reasonable to assume that with a given rate of cooling, the lattice constant of alpha martensite in a given steel has a definite value. The beta martensite of a given steel must also have a definite cubic lattice with carbon atoms within it, and therefore the lattice should have a definite value, as is actually confirmed. Thus, Dr. Dowdell's assertion as to the inconstancy of the parameters of alpha and beta martensites, is certainly incorrect.

Lastly it may be remarked that Dr. Sauveur and Dr. Dowdell approve my view regarding the decomposition of austenite during cooling through the  $A_1$  point, that is, they accept the identity

<sup>11</sup>Science Reports, Tohoku Imperial University, Vol. 17, 1928, p. 43.

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$A_1$  transformation = (1)  $\rightarrow$  (2)

as correct. It then follows logically:—

$A_1$  transformation = austenite  $\rightarrow$  martensite  $\rightarrow$  pearlite,

that is, the  $A_1$  change takes place always through the intermediate stage—martensite—irrespective of the velocity of cooling. But it is curious to see Dr. Dowdell's remarks: "It does not seem that pearlite can form direct from martensite, because martensite will not be formed at the high temperature of 700 degrees Cent. where the steel is plastic." Probably he does not realize the exact meaning of the above identity, or does not take the martensite as I have defined.

Dowdell's remark "martensite is formed at a temperature of about 300 degrees Cent. or below," really signifies that the martensite formed at the low temperature cannot easily pass on to pearlite, and remains as such at room temperature, while the martensite formed at a high temperature rapidly changes into pearlite.

According to Dowdell, stress is a necessary factor in the formation of martensite, but this view is probably not correct. Obviously, the stress is developed during martensitization; but it is the effect and not the cause. For, in the case of quenching a very thin wire or ribbon in a current of hydrogen gas, the cooling or hardening is nearly uniform throughout the specimen and hence the stress will be very small; but the specimen is equally well hardened, so that it may be concluded that the stress is not necessary for martensitization.

As to Dr. W. L. Fink's discussion, the following points deserve notice:

(1) In my theory of hardening, both alpha and beta martensites are responsible for the changes brought about by tempering, especially the latter martensite but not the alpha martensite alone, as considered by Dr. Fink. Hence some disagreements among the results as regards alpha martensite obtained by different investigators does not affect my theory.

(2) My reasoning why alpha martensite is an intermediate stage between austenite and beta martensite, is based on the mechanism of the atomic change from austenite to beta martensite; whether its physical property lies between those of austenite and martensite is not an important factor.

(3) The contention that the aging effect, which, Dr. Fink considers, cannot be explained by my theory, is easily realized from this theory. The expression such as "alpha martensite changes into beta martensite at 150 degrees Cent.," or "beta martensite decomposes at 270 degrees" means in reality that the velocity of the transformation, or that of decomposition is greatest in this range of temperatures. In both cases, the change occurs even at room temperature but its velocity is extremely small; as the temperature is raised, its velocity is increased at first very slowly and then more and more rapidly, reaching the great velocity at the respective temperature above referred to.

If from the quenched specimen, the outer layer containing alpha

martensite be ground off, the change alpha to beta martensite during tempering vanishes, but beta martensite slowly decomposes; hence the age hardening vanishes, but there remains only the aging effect of softening.

The fact that upon aging a hardened steel at 50 degrees Cent., the hardness increases to a maximum and then decreases, is a natural consequence from my theory. For, during a prolonged heating, alpha martensite changes into beta martensite accompanied by an increase of hardness, and at the same time, beta martensite also decomposes into iron and cementite, though by a much slower velocity, which will cause a decrease of hardness. The combined effect of these two must result in a maximum in the hardness-time curve during tempering, as actually observed.

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## THE APPLICATION OF SCIENCE TO THE STEEL INDUSTRY—SECTION V

BY DR. W. H. HATFIELD

### *Abstract*

*There is an increasing call for data regarding steel to be used at high temperatures and pressures. The author gives a review of the work done in determining the properties which various steels possess at elevated temperatures. The effect of temperature on tensile strength, fatigue properties, impact tests and the effect of corrosion at higher temperatures are discussed. A lengthy discussion and much data on various heat resisting steels are given.*

*The effect of steam at the higher temperatures is considered. Many of the high-temperature tests were conducted in atmospheres of gases and gas mixtures which approximate conditions under which materials are commonly used in industry.*

### EFFECT OF TEMPERATURE UPON STEELS

(Including a Treatment of Heat Resisting Steels)

THE development of higher pressures and temperatures in steam production, the call for steels having high strength and chemical resistance at elevated temperatures in the chemical industries, the general problem of exhaust valves and other demands, have made the study of the characteristics of steels at high temperatures of first importance. If the designer is to devise plant and machinery which can be safely operated at higher temperatures under diverse influences, the effect of those conditions upon the characteristics of the steel must be completely determined, and the designer given such concrete and reliable information that he may be able to apply the necessary factors of safety. The call for such data is imperative. Both in America and in Europe, capable investigators are attempting to explore this field successfully, but it is quite clear that some time must elapse before the data can be

This paper is the fifth section of the third Edward De Mille Campbell Memorial Lecture presented before the tenth annual convention of the society, held in Philadelphia, October 8 to 12, 1928, which is being published in six sections in consecutive issues of TRANSACTIONS. The author, Dr. W. H. Hatfield, is director of the Brown-Firth Research Laboratories, Sheffield, England.

available. The delay is due to the well recognized difficulty of dealing with the time effect. Chevenard,<sup>130</sup> Dickenson,<sup>131</sup> Lea,<sup>132</sup> Tapsell and Bradley,<sup>133</sup> Brown,<sup>134</sup> Jasper,<sup>135</sup> White,<sup>136</sup> Kanter and Spring,<sup>137</sup> French, Cross and Peterson,<sup>138</sup> and others, by substantial contributions to our knowledge in this field, have given us a sound foundation upon which to build.

To recapitulate the position, and the author feels entitled to do so, owing to the contribution which the Brown-Firth research laboratories are endeavoring to make to the subject, it is proved that steels may be plastically deformed by stresses imposed over long periods, which would be without measurable effect when imposed over short periods. Clearly, the first problem is to determine the value of stress which can be borne indefinitely without plastic deformation.

In a recent paper,<sup>139</sup> the author has discussed this question, and has expressed, or rather restated, the view that steels must have an elastic range, i.e., a range of stress within which they behave as practically elastic bodies. This follows from the fact that they are crystalline aggregates, and it holds good for the crystalline phase irrespective of temperature, or, how would the atoms orientate themselves to the space lattice? It is quite clear that a stress of a certain magnitude must be applied to destroy the lattice, otherwise it will persist. At low temperatures, the order of stress required is relatively high; at high temperatures, it is only nominal. When cohesion is overcome, plastic deformation results. Our difficulty today lies in the severe limitation as re-

<sup>130</sup>Chevenard, *Comptes Rendus*, 1919, 169, pp. 712 and 1922, 175, p. 486.

<sup>131</sup>Dickenson, *Journal*, Iron and Steel Institute, 1922, 106, p. 103, and *Engineering*, 1922, 114, p. 326.

<sup>132</sup>Lea, *Proceedings*, Institute of Mechanical Engineers, Vol. II, 1924, p. 1053.

<sup>133</sup>Tapsell and Bradley, *Engineering*, Vol. 120, 1925, p. 614, *et seq.*

<sup>134</sup>Brown, *Journal*, Institute of Metals, 1925, p. 21, and *Engineering*, Vol. 120, 1925, p. 461.

<sup>135</sup>Jasper, *Proceedings*, American Society for Testing Materials, Vol. 25 (part 2), 1925, p. 27.

<sup>136</sup>A. E. White, "Properties of Iron and Steel at High Temperatures," *TRANSACTIONS*, American Society for Steel Treating, Vol. 2, 1920, p. 521, and *Mechanical Engineering*, Vol. 49, 1927, p. 1093.

<sup>137</sup>Kanter and Spring, *Proceedings*, American Society for Testing Materials, June, 1928.

<sup>138</sup>French, Cross and Peterson, Bureau of Standards Technical paper No. 362.

<sup>139</sup>W. H. Hatfield, *Journal*, Iron and Steel Institute, Vol. I, 1928.

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gards sensitivity of measurement. In our own researches we work to a direct observation of  $1/40,000$  of an inch, others claim to be working to a direct observation of  $1/1,000,000$  of an inch, but it will be quite clear that deformation may, under certain conditions, be proceeding at a rate which is so slow as not to be observed, even with the most delicate measurement, except over the longest periods of time. This fact gives an opening for the expression of the abstract idea that "any metal will be deformed under any stress, given a long enough period of time." It is one of those propositions difficult to establish, and equally difficult to disprove. Nevertheless, on general grounds, as indicated above, the author holds the contrary view that steels have elastic ranges over the range of temperature in which the engineering world is interested, and, further, that it is our business to know what those ranges are. To leave theoretical discussion, the observation may be made that the point at issue is whether we can, or cannot, give the engineer a factor of safety as regards the time effect in the expectation of life of the part concerned. The answer is in the affirmative. From this point of view, it is interesting to consider the results of our present methods of attack.

In Britain there is a completely free exchange of ideas and a general collaboration between those responsible for design, and those engaged in the production of the material, and the subject is being followed from point to point with the greatest interest, in the hopes of immediate practical solution. The author has heard the view expressed by turbine engineers, that they would be in the position to design adequately with resultant suitable factors of safety, if they were in possession of the data concerning the stresses below which the rate of creep would not exceed  $0.000001$  inch per hour.

An extremely valuable summing up of the present position as regards the properties of steels for use in connection with superheated steam, has recently been given by R. G. C. Batson,<sup>140</sup> and he gives typical data concerning the limiting creep stress, as disclosed in Tables XX and XXI.

These data relate to mild carbon steels, and Batson states that even by increasing the strength of the steel by increasing the carbon content, a limiting creep stress exceeding 11,200 pounds per

<sup>140</sup>R. G. C. Batson, Institution of Civil Engineers, Engineering Conference, June, 1928.

square inch (5 tons per square inch) cannot be obtained, even with a carbon content of 0.51 per cent at 932 degrees Fahr. (500 degrees Cent.).

The limiting creep stresses stated in this paper are the result of work at the National Physical Laboratory.<sup>141</sup>

The limiting creep stress as determined at the National Physical Laboratory, is the stress above which there is a steady continuous creep of a measurable rate within the accuracy of their

Table XX  
Steel for Boiler and Superheater Tubes

		0.10 per cent Carbon Steel						0.17 per cent Carbon Steel					
Temperature		Limit of Proportionality		Ultimate Tensile Stress		Limiting Creep Stress		Limit of Proportionality		Ultimate Tensile Stress		Limiting Creep Stress	
Degrees Cent.	Degrees Fahr.	Tons per sq. in.	Lbs. per sq. in.	Tons per sq. in.	Lbs. per sq. in.	Tons per sq. in.	Lbs. per sq. in.	Tons per sq. in.	Lbs. per sq. in.	Tons per sq. in.	Lbs. per sq. in.	Tons per sq. in.	Lbs. per sq. in.
400	752	..	.....	..	.....	..	.....	6.5	14560	29.0	65000	13.4	30020
450	842	..	.....	..	.....	..	.....	5.5	12320	24.5	54900	8.5	19040
480	896	4.0	8960	17.0	38100	5.5	12320	5.0	11200	21.5	48200	6.0	13440
500	932	3.3	7390	14.9	33380	3.5	7840	4.8	10750	19.4	43500	4.8	10750
525	977	2.9	6500	13.7	30690	2.0	4480	4.4	9860	17.0	38100	3.5	7840
550	1022	2.5	5600	12.1	27100	1.3	2910	4.0	8960	15.0	33600	2.4	5380
570	1058	2.2	4930	10.5	23520	1.0	2240	3.5	7840	13.0	29100	2.0	4480
600	1112	1.9	4260	8.4	18820	0.6	1340	2.8	6270	11.0	24600	1.2	2690

measurements. Test pieces are loaded for several months at different values of stress, and the rate of creep to which the pieces settle down is determined. From these rates of creep a curve is drawn from which a close estimate is made of the limiting creep stresses. In a recent discussion, the author elicited the fact from Mr. Tapsell, who had been conducting this research, that four to five months were required to obtain a satisfactory value of limiting creep stress at one temperature. Quite clearly, therefore, researches having as their object the improvement of steel as regards an increase in the limiting creep stress, are severely handicapped owing to the great time required for discriminating examination of steels of varying composition and diverse heat treatments. This unsatisfactory position has been dealt with in the Brown-Firth research laboratories by arbitrarily deciding the conditions of the standardized test likely to be helpful in this direction. The meth-

<sup>141</sup>Special Report No. 2 of the Department of Scientific and Industrial Research, Engineering Research, National Physics Laboratory, London, 1927.

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**Table XXI**  
**0.23 Per Cent Carbon Steel for Superheater and Steam Drums**

Temperature		Limit of Proportionality		Ultimate Tensile Stress		Limiting Creep Stress	
Deg. Cent.	Deg. Fahr.	tons/sq. in.	lbs./sq. in.	tons/sq. in.	lbs./sq. in.	tons/sq. in.	lbs./sq. in.
400	752	6.3	14110	26.0	58240	13.5	30240
470	878	5.3	11870	18.5	41440	7.0	15680
530	986	4.0	9000	16.4	36740	4.0	8960
(approx.)							

od adopted consists in discovering, by static loading, the stress within which, at the temperature, stability of dimensions is attained within a period of 24 hours for a further period of 48 hours with an extension not exceeding the elastic deformation by 0.5 per cent on the gage length, with limitation as regards measurement for permanence of dimensions to 0.01 per cent of the gage length. This value the author has named the time-yield. It will be seen that a single value for a given temperature is obtained by this method in three days, and the matter is, therefore, brought, in point of time, within limits which permit a large number of tests being made in a reasonable period.

It is interesting to consider the value of the results so obtained in relation to the limiting creep stress values given by Batson as a result of the standard creep test determination. If we take mild carbon steel of about 0.25 per cent carbon, having a maximum stress value at ordinary temperatures of 72,000 to 74,000 pounds per square inch (32 to 33 tons per square inch), at 752 degrees Fahr. (400 degrees Cent.) the limiting creep stress value obtained is given at 30,240 pounds (13.5 tons) per square inch, whereas at this temperature the time-yield determination, as just discussed, gives a value closely approaching 26,880 pounds (12 tons) per square inch. At 932 degrees Fahr. (500 degrees Cent.) the limiting creep stress may be safely deduced as being about 13,400 pounds per square inch (6 tons per square inch) and the time-yield value obtained will be in the neighborhood of this figure. If one considers the steel at 1112 degrees Fahr. (600 degrees Cent.), it will be seen that the limiting creep stress is about 3360 pounds

per square inch (1.5 tons per square inch), and the time-yield value obtained is slightly under this figure. It thus becomes clear that if a comparison be made between the values obtained under the ordinary tensile test at elevated temperatures (see Table XXIII) the time-yield values are very materially below the yield values under the ordinary tensile test, and that they do, indeed, practically give a value which has more bearing upon design than the limiting creep figures obtained from the creep test method of research, since the latter does not base the final figures on initial extension.

The Table XXII is given in pounds per square inch (in tons per square inch) for a 0.27 per cent carbon steel the limiting creep stress, as determined by the National Physical Laboratory, the time-yield as determined by the Brown-Firth research laboratories, together with the limit of proportionality, the ordinary yield point and the maximum stress as regards the range of temperature extending from 572 to 1112 degrees Fahr. (300 to 600 degrees Cent.). It should also be remembered that the creep limit gives no indication of initial deformation, which is of great importance to the designer, whereas the time-yield does give such information.

As indicating the effect of various loads on the manner of deformation at a given temperature, Fig. 17 is of interest. In that graph will be found portrayed the resultant curves due to loading 0.40 per cent carbon steel at 752 degrees Fahr. (400 degrees Cent.) under 10, 12, 13, 14 and 15.3 tons per square inch respectively. This graph clearly and definitely demonstrates that the time-yield, as herein suggested and determined, is a value which

Table XXII

Temperature		Limit of Proportionality		Yield Point		Maximum Stress		Time-Yield		Creep Limit	
Deg. Cent.	Deg. Fahr.	Tons per sq. in.	Lbs. per sq. in.	Tons per sq. in.	Lbs. per sq. in.	Tons per sq. in.	Lbs. per sq. in.	Tons per sq. in.	Lbs. per sq. in.	Tons per sq. in.	Lbs. per sq. in.
300	572	13.2	29600	18.0	40300	40.5	90700	about 16.0	30800	29.0	65000
400	752	9.4	21060	15.8	35400	36.2	81100	about 11.0/12.0	26900	13.8	30900
500	932	6.0	13440	13.7	30700	26.1	58500	about 6.0	13400	5.0/6.0	11200/13400
600	1112	3.5	7840	...	.....	13.7	30700	1.0/1.5	2240/3360	1.5 about	3360

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**Table XXIII**  
**Tensile Tests on Carbon Steels at Elevated Temperatures**

0.14 Per Cent Carbon Steel									
Temperature, Deg. Cent.	Temperature, Deg. Fahr.	Limit of Proportionality, tons per square inch	pounds per square inch	Yield Point, tons per square inch	pounds per square inch	Maximum Stress, tons per square inch	pounds per square inch	Elongation, per cent	Reduction of Area, per cent
15	59	17.8	39870	17.9	40090	27.2	60930	38.0	61.0
100	212	15.8	35390	17.7	39650	28.1	63940	34.5	60.0
200	392	14.9	33380	17.3	38750	30.0	67200	27.0	58.0
300	572	13.1	29340	15.5	34720	32.5	72800	23.2	55.8
400	752	7.8	17470	12.9	28900	25.2	56450	37.6	68.5
500	932	.....	.....	.....	.....	20.0	44800	38.0	73.0
600	1112	.....	.....	.....	.....	12.5	28000	48.0	86.0
700	1292	.....	.....	.....	.....	6.8	15230	56.0	93.0
800	1472	.....	.....	.....	.....	4.0	8960	65.0	96.0
900	1652	.....	.....	.....	.....	2.0	4480	75.0	98.0
1000	1832	.....	.....	.....	.....	.....	.....	.....	.....
0.30 Per Cent Carbon Steel									
15	59	21.6	48380	22.2	53720	36.6	81980	26.6	50.3
100	212	18.6	41660	18.75	42000	33.7	75490	23.2	54.7
200	392	16.4	36740	20.7	46370	40.7	91170	22.4	46.8
300	572	14.3	32030	19.4	43400	43.55	91550	18.7	35.5
400	752	10.1	22620	19.4	43460	40.7	91170	28.0	50.6
500	932	.....	.....	15.0	33600	28.6	64060	30.4	65.8
600	1112	.....	.....	.....	.....	15.5	34720	51.0	80.0
700	1292	.....	.....	.....	.....	8.07	18080	48.8	86.0
800	1472	.....	.....	.....	.....	5.47	12250	60.7	96.0
900	1652	.....	.....	.....	.....	3.75	8400	68.0	97.0
1000	1832	.....	.....	.....	.....	2.65	5930	78.0	99.0
0.42 Per Cent Carbon Steel									
15	59	25.0	56000	26.6	59580	41.0	91840	22.4	47.0
100	212	20.5	45920	23.2	55960	40.08	89780	23.2	49.0
200	392	20.0	44800	25.0	56000	43.68	97840	17.6	42.0
300	572	20.8	46590	25.87	56950	49.67	111260	13.6	23.4
400	752	15.5	34720	25.2	56450	42.0	914080	28.8	58.0
500	932	10.0	22400	20.7	46370	33.5	75040	25.1	58.2
600	1112	.....	.....	.....	.....	21.0	47040	32.0	74.6
700	1292	.....	.....	.....	.....	10.1	22620	44.8	48.0
800	1472	.....	.....	.....	.....	5.95	13330	66.4	96.0
900	1652	.....	.....	.....	.....	4.17	9340	56.7	90.0
1000	1832	.....	.....	.....	.....	2.42	5420	80.0	99.0

can undoubtedly form a safe basis for design; particularly having in mind that the period over which observation is made, taken in conjunction with the degree of accuracy of measurement, discloses movements of a less order than one-millionth of an inch per inch per hour.

A good wrought iron is the commercial approximation to pure iron, and it is of interest to record that the time-yield values are very materially lower, as would be expected, than in the case of mild steel. A value of 14780 pounds per square inch (6.6 tons

per square inch) is obtained at 752 degrees Fahr. (400 degrees Cent.), while at 932 degrees Fahr. (500 degrees Cent.) this value is lowered to 8960 pounds per square inch (4.0 tons per square inch).

When considering the hardened and tempered alloy steels, one finds the values disappointing. For instance, 3 per cent nickel steel, which has a time-yield of 50,400 pounds per square inch (22.5

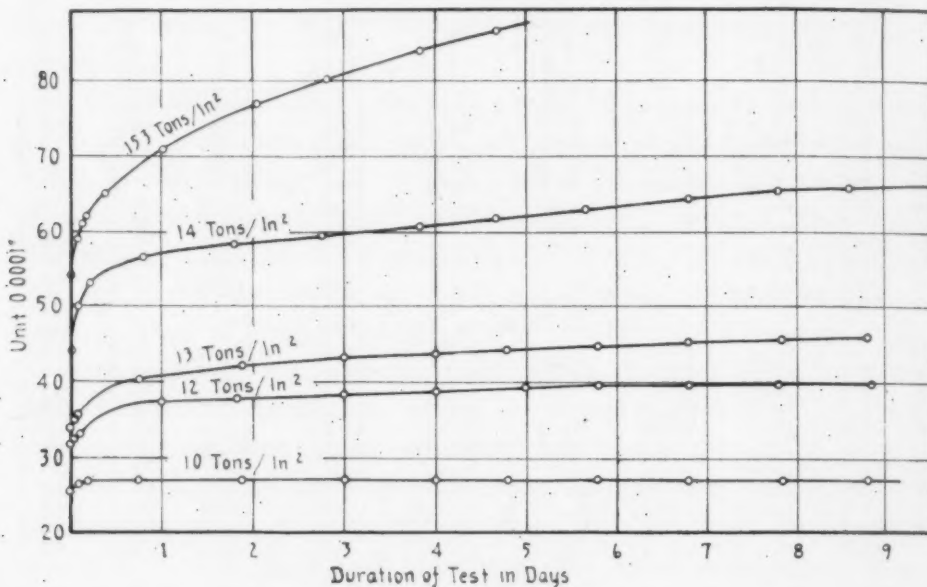


Fig. 17—Curves Showing the Effect of Various Loads on the Manner of Deformation at a Given Temperature.

tons) at 752 degrees Fahr. (400 degrees Cent.), discloses a value so low as 19,040 pounds per square inch (8.5 tons) at 932 degrees Fahr. (500 degrees Cent.): while a 60 to 65-ton hardened and tempered nickel-chromium-molybdenum steel at 932 degrees Fahr. (500 degrees Cent.) gives a value of 17,900 to 20,100 pounds per square inch (8.0 to 9.0 tons per square inch), which falls to 4480 pounds per square inch (2 tons per square inch) at 1112 degrees Fahr. (600 degrees Cent.). Tables XXIV and XXV are included as indicating the values of the ordinary tensile test as modified by increasing temperature, as regards the typical special steels. The nickel-chromium-tungsten steel given in Table XXV belongs to the new type of heat resisting steel which is now being developed, and it is of great interest to record that greatly increased values as regards time-yield determinations are obtained. This partic-

Tensi  
Elevated

Temperature, Deg. Cent.	Temperature, Deg. Fahr.
15	59
100	212
200	392
300	572
400	752
500	932
600	1112
700	1292
800	1472
900	1652
1000	1832

15	59
100	212
200	392
300	572
400	752
500	932
600	1112
700	1292
800	1472
900	1652
1000	1832

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Table XXIV

Tensile Tests on Hardened and Tempered High-Tensile Alloy Steels at Elevated Temperatures

Heat Treated 3 Per Cent Ni Steel									
Temperature, Deg. Cent.	Temperature, Deg. Fahr.	Limit of Proportionality, tons per square inch	pounds per square inch	Yield Point, tons per square inch	pounds per square inch	Maximum Stress, tons per square inch	pounds per square inch	Elongation, per cent	Reduction of Area, per cent
15	59	29.3	65630	32.2	72130	50.7	113570	23.3	59.2
100	212	28.0	62720	34.4	77060	48.5	108640	25.0	56.4
200	392	26.3	58910	34.0	76160	48.4	108420	25.4	56.0
300	572	25.2	56450	33.6	75260	47.7	106850	21.0	57.0
400	752	22.7	50850	33.3	74590	46.95	105170	21.6	50.4
500	932	.....	.....	22.4	50180	29.1	65180	28.0	79.4
600	1112	.....	.....	11.75	26320	18.67	41820	40.0	89.8
700	1292	.....	.....	.....	.....	11.5	25760	48.0	96.0
800	1472	.....	.....	.....	.....	6.53	14630	71.2	93.0
900	1652	.....	.....	.....	.....	4.025	9020	56.0	83.0
1000	1832	.....	.....	.....	.....	2.45	5490	72.6	99.0
Heat Treated Cr-Va Steel									
15	59	53.53	119910	55.55	124430	61.0	136640	21.6	61.5
100	212	53.9	120740	57.5	128800	62.6	140220	20.0	63.0
200	392	49.5	110880	53.12	118980	59.6	133500	15.6	61.5
300	572	43.5	97440	49.2	110210	55.6	124540	20.0	59.0
400	752	36.1	80860	46.6	104380	56.3	126110	15.6	61.5
500	932	15.5	34720	27.0	60480	32.6	73020	26.4	74.4
600	1112	.....	.....	.....	.....	24.8	55550	30.4	87.0
700	1292	.....	.....	.....	.....	11.6	25980	52.0	92.0
800	1472	.....	.....	.....	.....	6.7	15010	68.0	95.0
900	1652	.....	.....	.....	.....	4.7	10530	81.5	96.0
1000	1832	.....	.....	.....	.....	3.1	6940	73.5	97.0
Heat Treated Ni-Cr Steel									
15	59	46.5	104160	48.8	109310	57.8	129470	22.6	63.6
100	212	34.4	77060	47.5	106400	56.4	126330	23.2	67.7
200	392	32.6	73020	45.6	102140	54.85	122860	20.0	64.6
300	572	37.0	82880	48.5	108640	55.28	123830	20.0	50.2
400	752	30.4	68100	46.89	105340	55.41	124120	17.6	44.0
500	932	22.0	49280	38.97	87290	44.91	100600	21.6	66.5
600	1112	.....	.....	.....	.....	23.62	52900	32.0	88.0
700	1292	.....	.....	.....	.....	12.31	27570	42.4	90.0
800	1472	.....	.....	.....	.....	5.6	12540	74.3	92.0
900	1652	.....	.....	.....	.....	4.53	10150	52.0	84.6
1000	1832	.....	.....	.....	.....	2.97	6650	82.5	99.0

ular steel has a time-yield of 40,300 pounds per square inch (18.0 tons per square inch) at 932 degrees Fahr. (500 degrees Cent.); at 1112 degrees Fahr. (600 degrees Cent.) a value so high as 28,200 pounds per square inch (12.6 tons per square inch) is maintained. The author considers that sufficient data have been quoted concerning the time-temperature effect as regards ordinary steels, and he cannot too strongly recommend a study of the time-yield upon the lines indicated. In the Laboratories under his charge are

**Table XXV**  
**Tensile Tests on a Heat Treated Chromium Steel and Silico-chromium and**  
**Nickel-chromium-tungsten Steels at Elevated Temperatures**

Heat Treated 14 Per Cent Cr Steel									
Temperature, Deg. Cent.	Temperature, Deg. Fahr.	Limit of Proportionality, tons per square inch	Yield Point, tons per square inch	Maximum Stress, tons per square inch	Elongation, per cent	Reduction of Area, per cent			
15	59	28.6	64060	35.7	79970	48.5	108640	28.0	62.5
100	212	28.0	62720	35.8	80190	45.9	102820	24.8	65.4
200	392	28.5	63840	35.4	79300	43.5	97440	23.2	65.4
300	572	26.1	58460	33.9	75940	41.3	92510	21.0	64.0
400	752	23.4	52420	31.2	69890	37.8	84670	20.8	62.0
500	932	.....	.....	.....	.....	32.7	73250	16.8	56.6
600	1112	.....	.....	.....	.....	18.8	42110	20.5	67.0
700	1292	.....	.....	.....	.....	12.1	27100	36.8	85.4
800	1472	.....	.....	.....	.....	5.04	11290	45.6	91.5
900	1652	.....	.....	.....	.....	.....	.....	.....	.....
1000	1832	.....	.....	.....	.....	.....	.....	.....	.....
Si-Cr Steel									
15	59	48.5	108640	59.0	132160	68.0	152320	20.0	49.0
100	212	.....	.....	58.5	131040	68.5	153440	21.6	50.0
200	392	.....	.....	50.2	112450	63.0	141120	21.6	52.0
300	572	.....	.....	45.9	102820	58.2	130370	20.8	50.0
400	752	.....	.....	45.0	100800	57.0	127680	22.4	51.0
500	932	.....	.....	39.0	87360	45.3	101470	28.0	54.5
600	1112	.....	.....	.....	.....	24.5	54880	47.0	87.0
700	1292	.....	.....	.....	.....	10.08	22580	61.5	95.0
800	1472	.....	.....	.....	.....	4.87	10910	60.3	97.0
900	1652	.....	.....	.....	.....	2.82	6320	88.0	97.0
1000	1832	.....	.....	.....	.....	.....	.....	.....	.....
Ni-Cr-W Steel									
15	59	23.2	51970	36.0	80640	63.0	141,120	23.5	24.0
100	212	.....	.....	37.5	84000	53.5	119840	24.8	33.2
200	392	.....	.....	35.3	79070	49.4	110660	20.8	28.6
300	572	.....	.....	32.6	73020	43.4	97220	22.5	32.8
400	752	.....	.....	29.4	65860	42.7	95650	20.8	32.8
500	932	.....	.....	26.0	58240	39.5	88480	23.5	46.0
600	1112	.....	.....	.....	.....	34.1	76380	25.0	55.4
700	1292	.....	.....	.....	.....	28.2	63170	28.0	61.0
800	1472	.....	.....	.....	.....	17.2	38530	40.0	74.0
900	1652	.....	.....	.....	.....	12.0	26880	51.0	79.0
1000	1832	.....	.....	.....	.....	6.02	13450	70.0	81.0

a number of machines at work, and data are being steadily accumulated which cannot fail to be of great value.

#### *The Influence of Temperature on the Fatigue Properties of Steels*

Comparatively little work has been done on the fatigue properties of materials at elevated temperatures, and, as yet, no correlation has been achieved as regards the astonishing indications as to fatigue values at high temperatures, and the values as regards

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static stress required to produce plastic deformation. The first appears to be that by Professor F. C. Lea and H. P. Budgeon,<sup>142</sup> who found that the range of stress for zero mean stress, varied only slightly until comparatively high temperatures were reached. Thus a 0.14 per cent carbon steel had a fatigue range of  $\pm 33,600$  pounds per square inch ( $\pm 15$  tons per square inch) at 60 degrees Fahr. (15 degrees Cent.), the same at 392 degrees Fahr. (200 degrees Cent.), beyond which the range increased slightly to a maximum of  $\pm 40,300$  pounds per square inch ( $\pm 18$  tons per square inch) at nearly 750 degrees Fahr. (400 degrees Cent.). At still higher temperatures the fatigue range began to fall. For a nickel-chromium steel (3.25 per cent nickel, 0.6 per cent chromium, 0.35 per cent carbon) the range fell slightly from  $\pm 58,200$  pounds per square inch ( $\pm 26$  tons per square inch) at 60 degrees Fahr. (15 degrees Cent.) to  $\pm 44,800$  pounds per square inch ( $\pm 20$  tons per square inch) at 572 degrees Fahr. (300 degrees Cent.), then rose to a maximum of  $\pm 51,500$  pounds per square inch ( $\pm 23$  tons per square inch) at 752 degrees Fahr. (400 degrees Cent.) after which it fell slowly to  $\pm 38,100$  pounds per square inch ( $\pm 17$  tons per square inch) at 1400 degrees Fahr. (760 degrees Cent.). Similar results with slightly displaced maxima and minima, were obtained on a nickel-chromium steel containing 0.3 per cent carbon, 3.6 per cent nickel and 0.6 per cent chromium. The fatigue range above 752 degrees Fahr. (400 degrees Cent.) may be greater than the limiting creep stress at the same temperature. Thus it would appear to be possible, according to the data presented, to run specimens of the 0.14 per cent carbon steel at 932 degrees Fahr. (500 degrees Cent.) with a range of stress of more than 29,000 pounds per square inch (13 tons per square inch), and a mean stress of 33,600 pounds per square inch (15 tons per square inch), although the limiting creep stress at this temperature was only 8960 pounds per square inch (4 tons per square inch). The speed of reversal was 2000 per minute.

Moore and Jasper<sup>143</sup> have carried out fatigue tests on some metals at high temperatures under conditions of zero mean stress. Their results are collected in Table XXVI. The speed of revolution in the above tests was 1500 revolutions per minute.

<sup>142</sup>F. C. Lea and H. P. Budgen, *Engineering*, October 3, 1924, p. 500.

<sup>143</sup>Moore and Jasper, *University of Illinois Bulletin*, No. 152, 1925.

Table XXVI

Cyclops metal annealed				0.49% Carbon steel normalized				1.02% Carbon steel				S. A. E. steel 2340 O. Q. 1454° Fahr. T. 806° Fahr.				S. A. E. steel 2340 O. Q. 1454° Fahr. T. 1200° F.			
Temp.				Temp.				Temp.				Temp.				Temp.			
Deg. Cent.				Deg. Cent.				Deg. Cent.				Deg. Cent.				Deg. Cent.			
Fahr.				Fahr.				Fahr.				Fahr.				Fahr.			
Fatigue limit lbs./sq. in.				Fatigue limit lbs./sq. in.				Fatigue limit lbs./sq. in.				Fatigue limit lbs./sq. in.				Fatigue limit lbs./sq. in.			
21	280	70	56000	21	290	70	36000	21	310	70	105000	21	340	70	76000	21	464	70	74000
260	500	554	52000	290	554	590	39000	310	590	464	86000	240	340	464	82000	240	464	464	75000
470	878	806	46000	430	806	806	44000	430	806	644	80000	340	340	644	69000	340	644	644	65000
600	1112	878	37000	480	878	896	42000	480	896	878	75000	470	470	878	68000	470	878	878	62000
650	1202	1004	35000	555	1004	1031	34000	555	1031	1143	62000	540	540	1143	30000	540	1004	1004	55000
700	1292	1143	30000	620	1143	1202	24000	650	1202		30000				>				

Temp.  
of  
Test  
Air

100° C.  
(212° F.)

200° C.  
(392° F.)

300° C.  
(572° F.)

400° C.  
(752° F.)

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(932° F.)

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Table XXVII  
Fatigue Ranges Pounds Per Square Inch

Temp. of Test	Armco Iron	0.17% Carbon Steel			Carbon Steels			
					0.24%	0.51%	0.53%	
Air	±2600 ±11.9	±28000 ±12.5	44300 +22 to 0	66100 +29.5 to	32500 +14.5	±29800 ±13.3	±32900 ±14.7	±27300 ±12.2
100° C. (212° F.)	±24000 ±10.7	±27500 ±12.3	47000 +21 to 0	65000 +29 to	29100 +13	±27500 ±12.3	±31000 ±13.85	±28900 ±12.9
200° C. (392° F.)	±25500 ±11.4	±27500 ±12.3	51500 +23 to 0	> 762000 +34 to	26900 +12	±27300 ±12.3	±31100 ±13.9	..... .....
300° C. (572° F.)	±33400 ±14.9	±35800 ±16.0	67200 +30 to 0	85100 +38 to	49300 +22	±35800 ±16.0	±43200 ±19.3	±29100 ±13.0
400° C. (752° F.)	±28900 ±12.9	±37600 ±16.8	58200 +26 to 0	67200 +30 to	40300 +18	±38100 ±17.0	±43000 ±19.2	..... .....
500° C. (932° F.)	±12300 ±5.5	±26200 ±11.7	38100 +17 to 0	44800 +20 to	17900 + 8	±29100 ±13.0	±32000 ±14.3	±19500 ± 8.7
600° C. (1112° F.)	... ± 6.8	±15200 ± 6.8	.....	.....	.....	±16350 ± 7.3	±16100 ± 7.2	±15700 ± 7.0

Results of the same nature have been obtained by Tapsell and Clenshaw<sup>144</sup> and by Tapsell.<sup>145</sup> Their results are collected in Table XXVII. The speed of alternations in these tests was 2400 cycles per minute. The fatigue test results at high temperatures will undoubtedly be affected by the speed of revolution, but no data appear to have been obtained at present.

#### *Influence of Temperature on the Behavior of Steels Under the Impact Test*

The notched-bar impact test is much used and the question naturally arises as to how such values are affected by higher temperatures. Dealing with the matter as regards the range including the higher contemplated steam temperatures, it is agreed by experimentalists, that a material of high notched-bar impact value at normal temperature, generally retains a sufficiently satisfactory value through the range of temperature under consideration, while steels giving a low value at normal temperatures almost invariably give much increased values at the higher temperatures.

It has long been understood that steels pass through a range

<sup>144</sup>Tapsell and Clenshaw, "Properties of Materials at High Temperatures," Department of Scientific and Industrial Research, Parts I, II and III, 1927 and 1928.

<sup>145</sup>Tapsell, *Journal*, Iron and Steel Institute, May, 1928.

of temperature in which they are considered to possess the characteristic of blue brittleness.<sup>146</sup> The first systematic investigation of the subject appears to be that of Charpy,<sup>147</sup> who examined three carbon steels, a 1 per cent nickel steel, and a nickel-chromium steel. He found that the carbon steels all showed minimum brittleness (maximum toughness), in the neighborhood of 210 to 400 degrees Fahr. (100 to 200 degrees Cent.) and maximum brittleness between 750 to 930 degrees Fahr. (400 to 500 degrees Cent.). He recorded that only small variations occurred in the impact values of the two alloy steels between 176 and 1112 degrees Fahr. (80 and 600 degrees Cent.). Guillet and Revillon,<sup>148</sup> carried out impact tests on ordinary and special steels at high temperatures, and also found that for ordinary steels there was maximum resilience between 230 and 320 degrees Fahr. (110 and 160 degrees Cent.), and a minimum between 750 and 840 degrees Fahr. (400 and 450 degrees Cent.). Martensitic steels gradually increased in resistance to impact up to 986 degrees Fahr. (530 degrees Cent.), but appeared to pass through a maximum at about 300 to 340 degrees Fahr. (150 to 170 degrees Cent.). Austenitic steel gradually decreased in resistance to notched-bar impact as the temperature rose. Goerens and Hartel<sup>149</sup> found that a dead mild steel had low resilience at temperatures under normal. The impact value increased with temperature up to room temperatures and then fell to a minimum at 932 degrees Fahr. (500 degrees Cent.) reaching a well defined maximum at 1112 degrees Fahr. (600 degrees Cent.); after which it fell rapidly. Langenberg<sup>150</sup> carried out an extended investigation on the notch-bar impact test on materials at low and elevated temperatures. He found in all cases, that the resistance to impact increased with rise of temperature up to a maximum, after which the energy absorbed falls with further increase of temperature. In some cases there was a falling off of the resistance in the range 200 to 300 degrees Fahr. (93 to 150 degrees Cent.), which he suggests is due to the phenomenon of "blue-brittleness".

<sup>146</sup>Howe, "The Metallurgy of Steel," 1891.

<sup>147</sup>Charpy, *Bulletin, Societe des Ingenieurs Civils de France*, 1906, p. 563.

<sup>148</sup>Guillet and Revillon, *Revue de Metallurgie*, Vol. 6, 1909, p. 918.

<sup>149</sup>Goerens and Hartel, *Zeitschrift für Anorganische und Allgemeine Chemie*, Vol. 81, 1913, p. 130.

<sup>150</sup>F. C. Langenberg, Carnegie Scholarship Memoirs, Vol. XII, 1923, and "Behaviour of Metals Under Normal and Sub-Normal Temperatures," a paper presented before the Eastern Sectional Meeting of the American Society for Steel Treating, June 14 and 15, 1923.



Greaves and Jones<sup>151</sup> describe an investigation into the effect of temperature on the notched bar test. Two maxima were obtained in all cases except that of stainless steel. Using standard B. E. S. A. test pieces, the first maximum occurs at different temperatures between 0 and 480 degrees Fahr. (— 20 and 250 degrees Cent.) for different materials, the temperature at which the minimum impact figure is reached varies between 932 degrees Fahr. (500 degrees Cent.) for Armco iron and 1202 degrees Fahr. (650 degrees Cent.) for some heat treated alloy steels, and the second maximum occurs between 1200 and 1470 degrees Fahr. (650 and 800 degrees Cent.). The positions of the maxima and minima are altered considerably when the size of the notch is varied. For a fairly wide range of radius of the notch, the curve does not markedly change its form, but a change of radius from 10 to 200 millimeters causes a distinct alteration in the curve. The temperature of the first maximum for Armco iron is lowered from 392 degrees Fahr. (200 degrees Cent.) with notch sizes up to 1 millimeter, to — 112 degrees Fahr. (— 80 degrees Cent.) with an unnotched test piece. Cold work tends to raise the temperature at which the first maximum occurs. Evidence is given which attempts to connect the phenomenon of "blue-brittleness" with the minimum. Stainless steel showed a rapid rise in impact resistance up to 212 degrees Fahr. (100 degrees Cent.) and then increased very gradually up to 1832 degrees Fahr. (1000 degrees Cent.): no well defined maxima or minima were observed.

Greaves and Jones's paper also contains an excellent bibliography of the subject. It would appear, therefore, that apart from the general trend of one or two maxima, the curve showing impact variations with temperatures is so dependent on previous treatment, size, and type of notch, that each material and treatment should be examined separately. No complete generalization as to effect of composition can be made from published results.

Tables XXVIII and XXIX give some conception to the magnitude of the changes which take place in the notch bar impact figures at elevated temperatures. Langenberg—size of test piece 30 by 30 by 155 millimeters, notch 2 millimeters radius, 15 millimeter depth.

Greaves and Jones—size of test piece 10 by 10 by 60 milli-

<sup>151</sup>Greaves and Jones, *Journal*, Iron and Steel Institute, September, 1925.

Table XXVIII

Material	Condition	Room temp., ft.-lbs.	1st Max.		Impact		Figures Min.	2nd Max.		Impact
			Temp., deg. Cent.	Temp., deg. Fahr.	Temp., deg. Cent.	Temp., deg. Fahr.		Temp., deg. Cent.	Temp., deg. Fahr.	
2% Ni, 1% Cr, 0.2% C	Annealed	475	260	500	725	540	1000	410	...	...
	Hardened and temp. 260° C.	600	110	230	803	400	750	450	...	...
3.3% Ni, 0.3% C	Annealed	159	260	500	469	540	1000	217	...	...
	Hardened and temp. 600° C.	313	260	500	604	540	1000	311	...	...
1.9% Ni, 1% Cr, 0.36% C	Annealed	198	400	750	478	540	1000	289	...	...
	Hardened and temp. 600° C.	307	260	500	657	540	1000	396	...	...
0.9% Cr, 0.36% C	Annealed	31	180	356	401	540	1000	195	...	...
	Hardened and temp. 450° C.	430	80	176	865	108	225	599	260	500
0.83% C	Annealed	24	400	750	181	...	...	...	...	...
	Hardened and temp.	65	120	250	235	177	350	170	260	500
0.18% C	Annealed	61	120	250	868	540	1000	412	...	...
	Hardened and temp. 260° C.	893	108	225	1067	540	1000	434	...	...

meters, 45 degree V-notch 0.25 millimeter root radius, 2 millimeters depth.

#### *Corrosion at the Higher Steam Temperatures*

The question as to how far internal surfaces in contact with steam at the higher temperatures and pressures will become more susceptible to corrosion, has been raised, notably by Stuger and Bohnenblust,<sup>152</sup> and the author is of the opinion that corrosion must be indeed accelerated under such conditions. This matter must be thoroughly borne in mind and every step taken, particularly in eliminating oxygen from the feed. Experiments conducted in the author's laboratories, on several steels at 1000 degrees Fahr. (540 degrees Cent.), have shown definite attack over so short a period as 6 hours, in the case of ordinary steel when the steam was formed from water containing 5 cubic centimeters of oxygen per litre. Examples of the rustless and other steels were tested at the same time, and the results were as follows, see Table XXX, the figures being given as gain in weight in milligrams per square centimeter of surface.

<sup>152</sup>Stuger and Bohnenblust, Brown Boveri, November, 1927.

Mat  
Swedish  
Iron  
Armco I  
Armco I  
0.18%  
0.13%  
0.13%  
0.13%  
3.39%  
0.45%  
3.77%  
Carbon  
3.77%  
Carbon  
3.72%  
Cr, 0.23  
3.72%  
Cr, 0.23  
3.72%  
Cr, 0.23  
12.37%  
Stainless  
Carbon

temp

Table XXIX

Material	Condition	Room temp., kgm.	1st Max.			Impact Figures Min.			2nd Max.		
			Temp., Deg. Cent.	Temp., Deg. Fahr.	Impact, kgm.	Temp., Deg. Cent.	Temp., Deg. Fahr.	Impact, kgm.	Temp., Deg. Cent.	Temp., Deg. Fahr.	Impact, kgm.
Swedish Bar Iron	As rolled	16 & 4.5	15	60	16	540	1000	5	700	1290	10
Armco Iron A	As rolled	5	200	392	12	500	930	4	800	1470	5
Armco Iron B	As rolled	1	150	300	5	500	930	2.5	750	1380	4
0.18% Carbon	As rolled	15.5	15	60	15.5	600	1110	5	700	1290	21
0.13% Carbon	As rolled	2	100	212	3	600	1110	2	800	1470	5
0.13% Carbon	Annealed	1.5	100	212	3	600	1110	2	800	1470	5
0.13% Carbon	Water quenched	2.8	100	212	3	600	1110	1.5	800	1470	5
	1652° Fahr.	1.5	200	392	6	550	1020	2.5	750	1380	19
0.39% Carbon	Normalized	1.5	150	300	7.5	550	1020	3	750	1380	19
	Hardened and temp.										
	1200° Fahr.	1.0	200	392	3.5	550	1020	2	870	1600	11.5
0.45% Carbon	Normalized	2.0	150	300	11.0	550	1020	3.5	690	1270	25.2
	Hardened and temp.	5.5	250	482	6.5	600	1110	3	750	1380	15
3.77% Ni, 0.35% Carbon	Hardened and temp.										
	1200° Fahr.	2.5	250	482	5.5	550	1020	2.5	750	1380	13.5
3.77% Ni, 0.35% Carbon	Annealed	1.5	150	300	3	600	1110	2	800	1470	12.5
3.72% Ni, 0.92% Cr, 0.23% Carbon	O. H. and temp.	5.5	15	60	5.5	600	1110	3	800	1470	13
3.72% Ni, 0.92% Cr, 0.23% Carbon	1200° F.										
3.72% Ni, 0.92% Cr, 0.23% Carbon	W. Q.										
	O. H. and temp.	1.0	300	572	4.5	600	1110	3	800	1470	13
	1200° F.										
3.72% Ni, 0.92% Cr, 0.23% Carbon	S. C.										
	1652° F.	S. C. 3 Rises to 10 at 392° Fahr. and then gradually to 13.5 at 1832° Fahr.									
Stainless steel	O. H. and temp.	3 Rises to 7 at 212° Fahr. and then remains fairly constant up to 1290° Fahr., after which it rises to 13.5 at 1832° Fahr.									
12.37% Cr, 0.43% Carbon	1292° F.										

### Heat Resisting Steels

Heat resisting steels are essentially steels which will work in temperatures approximately ranging from 1100 to 1830 degrees

Table XXX

Material	Gain in Weight
0.25 per cent carbon steel	0.62
High tensile alloy steel	0.73
14 per cent chromium stainless steel	0.24
18 per cent chromium 8 per cent nickel, rustless steel	0.01

Fahr. (600 to 1000 degrees Cent.), with a much less scaling effect than occurs with ordinary ferrous materials. A number of interesting papers have been published dealing with the subject. Dickenson<sup>153</sup> studied the differential effect on several steels and alloys at various temperatures. McCormick<sup>154</sup> dealt with the effect of the furnace atmosphere in producing scale on steels of varying carbon content. There are several papers dealing essentially with valve steels; Aitchison<sup>155</sup> indicated the value of chromium steels, while Grard,<sup>156</sup> after an interesting description of various tests, claimed excellent characteristics for chromium-silicon steels. Mahoux<sup>157</sup> was responsible for an interesting metallurgical survey, and Henshaw<sup>158</sup> has made a review of the position to date. The latter author gave much valuable data concerning the composition and characteristics of heat-resisting steels in current use for meeting the onerous conditions imposed on aero-engine valves, and discussed the chromium, chromium-silicon, nickel-chromium, cobalt-chromium, and nickel-chromium-tungsten steels. The author has published two papers, one dealing with the resistance to scaling of such materials,<sup>159</sup> and the other dealing with the mechanical properties.<sup>160</sup>

The subject, as regards the mechanical properties of the materials, can be well considered, and indeed can only, at the present time, be studied by arbitrarily selecting a temperature and obtaining the comparative results obtained under a standardized tensile test. As will be clear from the foregoing matter, however, in this section, in which the question of the effect of the manner of applying the load upon the nature of the deformation, has been discussed, only those investigators who have given attention to the subject can appreciate how difficult it is to present comparative data concerning the strength of steel at high temperatures. In

<sup>153</sup>Dickenson, *Journal, Iron and Steel Institute*, No. II, 1922, p. 103.

<sup>154</sup>G. C. McCormick, "Furnace Atmospheres and Their Relation to the Formation of Scale," *TRANSACTIONS, American Society for Steel Treating*, Vol. 2, 1922, p. 1006.

<sup>155</sup>Aitchison, *Automobile Engineering*, Vol. IX, 1919, p. 401, and *Engineering*, Vol. CVIII, 1919, pp. 799-834.

<sup>156</sup>Grard, *Comptes Rendus*, Vol. CXXXXI, 1925, p. 1143.

<sup>157</sup>Mahoux, *Revue de Metallurgie, Memoires*, XXII, 1925, p. 39.

<sup>158</sup>Henshaw, Paper read before the Royal Aeronautical Society, December 2, 1926.

<sup>159</sup>W. H. Hatfield, *Journal, Iron and Steel Institute*, Vol. CXV, 1927.

<sup>160</sup>W. H. Hatfield, *Journal, Iron and Steel Institute*, 1928.



Table XXXI  
Tensile Tests at 15° and 800°C. (59° and 1472°F.)

Steel No.	Material	C	Mn	Si	P	S	Cr	Ni	Mo	V	Co	W	Al	Cu	Condition	15° C. (60° F.)				(1472° F.)			
																Y. P. tons, and lbs., per sq. in.	M. S. tons, and lbs., per sq. in.	Elongation, %	R/A, %	M. S. tons, and lbs., per sq. in.	Elongation, %	R/A, %	
1	Wrought Iron	0.09	0.08	0.10	....	....	....	....	....	....	....	....	....	....	As forged	13.78	22.34	29.0	48.43	2.03	53.5	58.0	
2	Mild Steel	0.14	0.56	0.02	0.048	0.04	....	....	....	....	....	....	....	....	Normalized	30900	50000	38.0	61.0	4.0	75.0	98.0	
3	3% Nickel	0.34	0.60	0.29	0.021	0.012	....	3.28	....	....	....	....	....	....	O. H. 850° C.	43700	60900	23.3	59.2	5.37	52.0	82.2	
4	3% Ni-Cr	0.29	0.60	0.27	0.014	0.018	0.89	3.41	....	....	....	....	....	....	T. 640° C.	72100	113600	24.0	63.6	5.60	74.3	78.0	
5	Ni-Cr Mo	0.30	0.57	0.24	0.010	0.015	0.61	2.38	0.59	....	....	....	....	....	O. H. 850° C.	109300	129500	24.0	66.0	7.68	70.3	90.0	
6	Cr-Va	0.43	0.59	0.22	0.039	0.034	1.14	0.32	....	0.18	....	....	....	....	T. 620° C.	126300	142700	18.7	62.0	6.70	68.0	95.0	
															O. H. 850° C.	50.9	58.5	15.2	23.4	8.8	48.0	83.0	
															T. 650° C.	114000	131000	....	....	....	....	....	
7	Si-Cr Valve	0.35	0.57	3.27	....	....	9.45	....	....	....	....	....	....	....	Annealed	52.0	70.0	13.0	20.5	6.05	66.5	99.0	
8	Si-Cr Valve	0.57	0.42	4.07	....	....	8.5	....	....	....	....	....	....	....	O. H. 1075° C.	116500	156800	20.0	49.0	3.93	81.6	99.3	
															T. 900° C.	59.0	68.0	....	....	8800	....	....	
9	High-Speed	0.55	0.24	0.23	....	....	3.36	....	....	0.47	....	14.75	....	....	A. C.	132200	152300	....	....	....	....	....	
10	High-Speed	0.56	0.20	0.21	....	....	4.15	....	....	0.65	....	16.05	....	....	Annealed	....	47.45	21.6	38.7	8.7	43.0	82.0	
11	High-Speed	0.54	0.22	0.39	....	....	4.08	....	....	0.84	....	18.0	....	....	Annealed	....	106300	22.0	38.5	19490	51.0	84.5	
12	High-Speed Super	0.58	0.12	0.41	....	....	3.90	....	7.91	0.99	5.41	0.36	....	....	Annealed	28.3	51.6	16.6	27.5	10.5	42.4	68.5	
																63400	115600	....	....	23520	....	....	
																48.8	57.1	15.2	23.4	8.8	48.0	83.0	
																109300	127900	....	....	19710	....	....	

H. R. Steel = Heat-Resisting Steel. O. H. = Oil Hardened. T. = Tempered. A. C. = Air Cooled. A. H. = Air Hardened. W. Q. = Water Quenched.

Table XXXI (Continued)  
Tensile Tests at 15° and 800°C. (59° and 1472°F.)

Steel No.	Material	C	Mn	Si	S	P	Cr	Ni	Mo	V	Co	W	Al	Cu	Condition	15° C. (60° F.) (1472° F.)				800° C.			
																Y. P. tons, and lbs., per sq. in.	M. S. tons, and lbs., per sq. in.	Elongation, %	R/A, %	Y. P. tons, and lbs., per sq. in.	M. S. tons, and lbs., per sq. in.	Elongation, %	R/A, %
13	Cobalt-Chrome	1.51	0.23	0.44	.....	13.22	1.68	.....	.....	.....	5.17	.....	.....	.....	Softened	81500	122300	13.0	16.5	81500	122300	13.0	.....
14	Valve Stain-less	0.07	0.10	0.10	.....	.....	14.32	.....	.....	.....	.....	.....	.....	.....	Fully tempered	18.0	31.0	40.0	74.26	3.05	77.6	99.8	.....
	Iron	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....	40300	69400	.....	.....	40300	69400	.....	.....
15	Experimental	0.09	0.39	0.37	.....	.....	18.52	0.26	.....	.....	.....	.....	.....	.....	A. C. 900° C.	19.0	28.6	16.0	.....	19.0	28.6	16.0	.....
16	Experimental	0.16	0.30	1.31	.....	.....	14.04	.....	.....	.....	.....	.....	.....	.....	A. H. 950° C.	42600	64100	.....	.....	42600	64100	.....	.....
17	Experimental	0.31	0.30	1.16	.....	.....	13.17	0.32	.....	.....	.....	.....	.....	.....	T. 820° C.	.....	83800	27.3	64.0	.....	83800	27.3	64.0
18	Stain-less	0.36	0.12	0.26	.....	.....	14.66	0.30	.....	.....	.....	.....	.....	.....	A. H. 950° C.	31.0	45.0	28.0	60.9	69400	100800	5.79	76.8
	Steel	0.54	0.33	0.94	.....	.....	14.56	0.88	.....	.....	.....	.....	.....	.....	T. 820° C.	69400	100800	22.0	45.0	69400	100800	12.97	83.3
19	Valve	0.23	0.18	0.10	.....	.....	Nil	12.23	.....	.....	.....	.....	.....	.....	O. H. 950° C.	.....	58.0	.....	.....	.....	58.0	.....	.....
20	Experimental	0.27	0.93	0.79	.....	.....	0.05	25.7	.....	.....	.....	.....	.....	.....	T. 600° C.	32.8	54.6	25.5	54.3	32.8	54.6	25.5	54.3
21	Experimental	0.23	0.35	0.12	.....	.....	Nil	36.5	.....	.....	.....	.....	.....	.....	As received	73500	122300	.....	.....	73500	122300	.....	.....
22	Experimental	0.09	0.25	0.03	.....	.....	Nil	11.91	.....	.....	.....	1.85	.....	.....	A. C. 950° C.	83.0	99.0	12.0	31.0	83.0	99.0	12.0	31.0
23	Experimental	0.11	0.34	0.21	.....	.....	14.84	10.16	.....	.....	.....	.....	.....	.....	A. C. 950° C.	185900	221760	.....	.....	185900	221760	.....	.....
24	Experimental	0.09	0.25	0.03	.....	.....	Nil	11.91	.....	.....	.....	1.85	.....	.....	A. C. 950° C.	13.1	40.0	52.0	70.4	13.1	40.0	52.0	70.4
	Water Quenched	0.23	0.35	0.12	.....	.....	Nil	36.5	.....	.....	.....	.....	.....	.....	A. C. 950° C.	29300	89600	.....	.....	29300	89600	.....	.....
25	Experimental	0.09	0.25	0.03	.....	.....	Nil	11.91	.....	.....	.....	1.85	.....	.....	A. C. 950° C.	22.8	40.4	44.5	64.0	22.8	40.4	44.5	64.0
26	Experimental	0.11	0.34	0.21	.....	.....	14.84	10.16	.....	.....	.....	.....	.....	.....	A. C. 950° C.	51100	90500	.....	.....	51100	90500	.....	.....
27	Experimental	0.09	0.25	0.03	.....	.....	Nil	11.91	.....	.....	.....	1.85	.....	.....	A. C. 950° C.	64.6	71.7	12.5	32.0	64.6	71.7	12.5	32.0
28	Experimental	0.11	0.34	0.21	.....	.....	14.84	10.16	.....	.....	.....	.....	.....	.....	1000° C.	144700	160600	.....	.....	144700	160600	.....	.....
29	Experimental	0.09	0.25	0.03	.....	.....	Nil	11.91	.....	.....	.....	1.85	.....	.....	As rolled	24.6	45.2	52.0	68.1	24.6	45.2	52.0	68.1
30	Experimental	0.11	0.34	0.21	.....	.....	14.84	10.16	.....	.....	.....	.....	.....	.....	As rolled	55100	101200	.....	.....	55100	101200	.....	.....

H. R. Steel = Heat-Resisting Steel. O. H. = Oil Hardened. T. = Tempered. A. C. = Air Cooled. A. H. = Air Hardened. W. Q. = Water Quenched.

Table XXXI (Continued)  
Tensile Tests at 15° and 800°C. (59° and 1472°F.)

Table XXXI (Continued)  
Tensile Tests at 15° and 800°C. (59° and 1472°F.)

Steel No.	Material	C	Mn	Si	S	P	Cr	Ni	Mo	V	Co	W	Al	Cu	Condition	15° C. (60° F.)				800° C. (1472° F.)			
																Y. P., tons, and lbs., per sq. in.	M. S., tons, and lbs., per sq. in.	Elongation, %	R/A, %	Elongation, %	M. S., tons, and lbs., per sq. in.	Elongation, %	R/A, %
25	Experi-mental	0.29	0.29	0.17	.....	.....	14.08	11.90	.....	.....	.....	.....	.....	.....	A. C.	18.2	42.8	64.0	65.0	12.65	30.4	33.0	
26	Experi-mental	0.41	0.46	1.10	.....	.....	13.94	9.84	.....	.....	3.15	.....	.....	.....	A. C.	40700	95900	9.0	12.6	13.10	37.0	68.5	
27	Experi-mental	0.34	0.33	1.18	.....	.....	13.80	8.44	1.44	.....	.....	.....	1.06	.....	A. C.	54200	103900	21.0	.....	14.33	68.0	.....	
28	Experi-mental	0.64	0.61	1.54	.....	.....	14.3	13.0	.....	.....	.....	1.0	.....	.....	A. C.	48200	125400	22.0	31.5	12.5	52.0	67.0	
29	Experi-mental	0.55	0.73	1.45	.....	.....	13.10	12.78	.....	.....	.....	2.3	.....	2.58	A. C.	79700	126300	22.5	32.0	14.0	43.0	65.3	
30	H. R. Steel	0.42	0.31	0.69	.....	.....	13.30	8.87	.....	.....	.....	3.42	.....	.....	A. C.	34.8	56.8	7.5	.....	22.75	16.0	32.8	
31	H. R. Steel	0.57	0.48	0.72	.....	.....	13.97	9.5	.....	.....	.....	3.36	.....	.....	As forged	40.1	62.9	29.0	33.0	22.50	24.0	47.0	
32	Experi-mental	0.76	0.69	0.76	.....	.....	12.10	9.38	.....	.....	.....	3.07	.....	.....	A. C.	89800	140900	11.0	11.0	15.4	54.0	71.2	
33	Experi-mental	0.61	0.67	0.4	.....	.....	14.94	10.36	.....	.....	1.91	5.94	.....	.....	1050° C.	63600	127700	12.5	12.0	13.65	33.6	66.4	
34	Experi-mental	0.47	0.70	0.14	.....	.....	12.47	19.67	.....	.....	.....	3.57	.....	.....	1000° C.	78800	116500	.....	.....	30580	19.30	18.4	37.8
35	Experi-mental	0.13	0.13	0.56	.....	.....	17.88	7.74	.....	.....	.....	.....	.....	.....	W. Q.	36.8	45.7	55.0	39.17	43250	7.48	10.4	14.5
36	Experi-mental	0.41	0.13	0.11	.....	.....	19.92	8.92	.....	.....	.....	.....	.....	.....	1200° C.	39200	106600	41.0	50.9	8.93	27.8	67.0	
37	Experi-mental	0.34	0.48	0.85	.....	.....	18.51	7.78	.....	.....	.....	3.09	.....	.....	W. Q.	24.0	53.35	10.0	.....	20000	21.3	.....	.....
38	H. R. Steel	0.38	0.68	1.21	.....	.....	18.72	9.74	.....	.....	.....	3.27	.....	.....	A. C.	35.2	70.0	27.5	28.9	477100	20.0	41.5	59.0
39	Experi-mental	0.30	0.52	1.46	.....	.....	17.70	7.0	.....	.....	.....	4.23	.....	.....	1000° C.	29.2	50.6	37.5	52.0	44800	17.8	38.0	53.0
															1000° C.	65400	113040	.....	.....	39870	.....	.....	.....

H. R. Steel = Heat-Resisting Steel. O. H. = Oil Hardened. T. = Tempered. A. C. = Air Cooled. A. H. = Air Hardened. W. Q. = Water Quenched.

Table XXXI (Continued)  
Tensile Tests at 15° and 800°C. (59° and 1472°F.)

Steel No.	Material	C	Mn	Si	S	P	Cr	Ni	Mo	V	Co	W	Al	Cu	Condition	15° C. (60° F.)				800° C. (1472° F.)									
																Y. P. tons, and lbs., per sq. in.	M. S. tons, and lbs., per sq. in.	Elongation, %	R/A, %	M. S. tons, and lbs., per sq. in.	Elongation, %	R/A, %							
40	Experi- mental	0.58	0.36	4.0	....	....	15.54	8.14	....	....	....	....	....	....	A. C. 1000° C.	70.0	156800	20.0	15.25	34160	68.0	71.8							
41	Experi- mental	0.59	0.27	1.08	....	....	19.10	7.73	9.73	....	....	....	....	....	Ann. 700° C.	47.7	60.0	4.5	4.0	20.98	36.0	49.6							
42	Experi- mental	0.24	0.23	0.23	....	....	19.74	Nil	....	....	....	2.91	....	....	A. C.	106900	134400	....	....	47000	....	....							
43	Experi- mental	0.25	0.42	3.24	....	....	19.4	0.25	....	....	....	3.25	....	....	Ann. 700° C.	31.6	41.8	10.0	12.0	4.47	7.20	94.0							
44	Experi- mental	0.06	0.39	0.39	....	....	23.22	24.02	....	....	....	....	....	....	Ann. 700° C.	70800	93600	2.0	4.0	6.02	73.5	96.0							
45	Experi- mental	0.19	0.32	0.84	....	....	20.29	19.29	....	....	....	....	....	....	A. C. 1050° C.	88700	92500	48.0	67.0	13480	33.5	38.4							
46	Experi- mental	0.25	0.96	2.20	....	....	23.85	17.21	....	....	....	....	....	....	A. C. 1050° C.	42.100	88700	35.0	49.0	29560	8.0	17.0							
47	H. R. Steel	0.29	1.16	1.81	....	....	14.95	24.82	....	....	....	2.86	....	....	As received	49.2	56.4	20.0	27.5	15.0	23.0	46.0							
48	Experi- mental	0.48	0.85	0.19	....	....	24.75	16.39	....	....	....	3.85	....	....	A. C. 1050° C.	24.2	47.7	35.5	50.0	19.58	24.0	51.0							
49	Experi- mental	0.52	1.01	1.41	....	....	23.90	22.29	....	....	....	3.13	....	....	A. C. 1050° C.	54200	106900	29.0	35.0	43860	60.0	57.7							
50	Experi- mental	0.35	1.36	0.21	....	....	10.9	35.15	....	....	....	....	....	....	As rolled	67200	108800	26.0	31.0	16.55	40.0	49.8							
51	Nichrome	0.06	0.89	0.71	....	....	11.69	60.40	....	....	....	....	....	....	As rolled	33.8	53.2	31.0	51.0	13.7	45.5	54.8							
52	Nichrome	0.81	1.50	0.58	....	....	12.8	66.0	....	....	....	....	....	....	As rolled	53800	103000	....	....	30686	40.0	38.5							
																33.03	54.6	26.0	30.0	12.25	40.0	38.5							
																74400	122300	....	....	27440	38.0	43.0							
																				34270									
H. R. Steel = Heat-Resisting Steel. O. H. = Oil Hardened. T. = Tempered. A. C. = Air-Cooled. A. H. = Air Hardened. W. Q. = Water Quenched.																													

H. R. Steel = Heat-Resisting Steel. O. H. = Oil Hardened. T. = Tempered. A. C. = Air-Cooled. A. H. = Air Hardened. W. Q. = Water Quenched.

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their own laboratory they have, however, compared a large number of steels and alloys of various compositions, at selected temperatures, and in Table XXXI the results of tensile tests at ordinary temperatures and at 1472 degrees Fahr. (800 degrees Cent.), together with the analyses and heat treatments, will be found.

The actual conditions of test were as follows:—The test pieces were 0.357 inch in diameter, with a parallel length of  $1\frac{1}{4}$  inches. The specimen was heated in situ in an electric resistance furnace. The time taken to obtain the temperature was three-quarters of an hour to one hour, and then the selected temperature of 1472 degrees Fahr. (800 degrees Cent.) was maintained for half an hour before the test piece was pulled. The speed of pulling was in all cases at the rate of a quarter of an inch elongation per minute. It is well appreciated that the results derived from a tensile test at high temperatures should be interpreted in a different manner from those obtained from such tests at ordinary temperatures; but when allowance is made in accordance with existing knowledge, such values sufficiently afford a means of comparison. A careful scrutiny of the data given in this table shows the extent to which, by modifying the composition of steel, its strength at high temperatures may be very materially improved.

It will be seen that wrought iron at 1472 degrees Fahr. (800 degrees Cent.) gives a value so low as 4550 pounds per square inch (2.03 tons per square inch), and this is of interest as a basis of comparison for the other materials. Mild carbon steel gives a value of 8960 pounds per square inch (4 tons per square inch), while the highest value given for any of the ordinary non-heat-resisting steels, can be taken to be under 17,900 pounds per square inch (8 tons per square inch). It will be seen that chromium, even when added up to 14 or 18 per cent, when the carbon is very low and in the absence of other elements, does not increase the strength at high temperatures. It will be seen that nickel alone, even when added in considerable quantities, does not result in very high values at this particular temperature. It will also be seen that if tungsten is added along with the chromium, or if this element is added along with the nickel, no outstanding result is obtained. It will, however, be clear that if tungsten is added in the presence of both nickel and chromium, and providing that a fair percentage of carbon is present, then maximum stress values

are obtained of 44,800 to 56,000 pounds per square inch (20 to 25 tons per square inch) as compared with 4,500 to 9,000 pounds per square inch (2 to 4 tons per square inch) for ordinary iron and steels.

This data, therefore, experimentally establishes the position of that very interesting class of chromium-nickel-tungsten steels which are now being used in increasing quantities.

Turning now to the discussion of the general problem dealt with, the author would like to be able to give satisfactory explanations of—

- (1) Why the characteristics of steel are so modified with increasing temperature.
- (2) Why the added elements modify the properties of the steel at the high temperature in the manner disclosed.

The whole of the explanation turns upon an understanding of the nature of cohesion, and that again awaits the more complete elucidation of the nature of the atom and the electron. Whether such an understanding will help to produce materials of exceptional characteristics is problematical, since, even when our knowledge of the atom and its fields of force is complete, we shall still have to await the next phase, which obviously consists in learning how to make use of that knowledge. Progress will probably take the same line as in the past, i.e., by experimentally determining, in a more or less empirical manner, the various factors which affect the cohesive forces. It is, of course, extremely desirable to know many things. For instance, what determines the space lattice with which the various atoms crystallize; why such different space lattices behave differently, as clearly shown by the figures herein stated; the causes of different solubilities; whether compounds can exist as such in solid solution; a visualization of the mechanism of the effect of the speed of application of stress.

As regards slip, the author has always been led to the conclusion that the smaller the crystals, the higher the resistance, owing to the multiplication of diverse orientations, whereas evidence would show that those steels have greatest strength at high temperature, which consist of very considerably sized crystals, thus leaving the resistance to slip to be mainly due to interatomic forces in their effect in influencing the intrinsic resistance of individual crystals to slip and rupture.

With regard to (1), when a tensile test piece is stressed, motion of the atoms in relation to one another takes place, with the resulting elastic deformation. After passing a certain value of stress, plastic deformation occurs. This is associated with slip along certain planes, and the movement is constrained to the degree to which the progressive balancing of the effort applied in overcoming the interatomic forces, has taken place. If the stress applied is high enough, deformation proceeds until rupture occurs. At higher temperatures the effort required to overcome the interatomic forces is less, owing to the modification in these forces with temperature. Under static loads, below the continuous creep limit, the test piece will elongate elastically at first, and then plastically, until the opposition, due to resistance to slip, balances the applied load. With greater loads the balance cannot occur, and the piece elongates until fracture takes place. Increasing temperature will decrease the resistance to slip, and hence "continuous creep" stress is lowered. With rapid loading, as in the case of a normal short-time tensile test, time is not allowed for the test piece to develop the amount of deformation which for the temperature and stress would otherwise occur. In consequence, the load required to fracture is greater than in the case of continuous loading. The difference between the two, increases with rise of temperature and with increase in the rate of application of the load. The influence of the reduction of area of cross-section during the test, must be considered as regards its influence in increasing the effective stress upon the material.

With regard to (2), in our actual present ignorance of the nature of slip, it is difficult to discuss the influence of the various elements introduced into the material. The author presumes—and indeed there is ample evidence that such is the case—that if the stress required to break the material at 1470 degrees Fahr. (800 degrees Cent.) under the ordinary tensile test is higher, then the elastic range and also the "continuous creep" stress are raised. Discussing, then, these general phenomena, it seems fair to postulate that the kind of atoms most likely to prevent slip, which is clearly the means by which improvements in these steels are made, would be those which, on account of the nature and magnitude of their fields of force, would cause most disturbance of the lattice. This would provide an influence operating at all temperatures, but

modified by the mobility of the atoms coincident with the temperature. Unfortunately, the strength at high temperature bears little relation to the strength at low temperature. Nickel is an element with a similar atom to iron, and crystallizes in the same manner at high temperatures; it would, therefore, not be expected to increase the strength much at such temperatures, when added to iron. It is interesting to note, as is well known, that a critical percentage of nickel increases the strength of iron at ordinary temperatures five fold; at 1470 degrees Fahr. (800 degrees Cent.) the effect of hardening is small. Chromium is an atom approximately similar to iron in size, and which crystallizes with a body-centered configuration. Reference to the experiments (Table XXXI) will show that neither at ordinary temperatures, nor at 1470 degrees Fahr. (800 degrees Cent.), does this element when added alone, even when present in large proportion, materially harden the steels, and yet it would appear to be of material help in providing great strength at high temperature. It might be deduced from the experiments, that nickel had some little influence in hardening up the iron, but that chromium had less. The two elements together certainly resulted in making it necessary to apply a stress several times greater at 1470 degrees Fahr. (800 degrees Cent.), than in the case of iron. The addition of tungsten in relatively small quantities in the presence of the elements nickel and chromium, without carbon, has not been determined. If however, the same amount of tungsten is added to iron alone, the result is negligible, and, incidentally, if the tungsten is added with only chromium present, such is also the case; likewise, if only nickel is present. One has, therefore, to consider that the effect to be explained is that tungsten added to iron in the presence of a sufficient quantity of nickel, chromium and carbon, produces this effect, yet, clearly, there is also an optimum carbon content. Now all that can be said is that tungsten is a heavier and larger atom, and the implication is that it should increase the strength of the iron; but why should it need a critical accompaniment of carbon, chromium, and nickel for strength at high temperatures, and why should its influence not be more marked at ordinary temperatures? It would appear that its influence is most effective when added to alloys in which the iron throughout the whole range of temperature exists in the gamma condition, the most successful of the steels being in the austenitic condition.

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*Scale Resistance of Heat Resisting Steels*

As regards the problem of throwing light upon the nature of the attack at high temperatures, it must at once be conceded that this is extremely difficult. It can be approached from the standpoint of corrosion at ordinary temperatures; and, indeed, in the literature of the subject, it is common to find the rustless and heat resisting steels dealt with together. The explanation of the phenomenon of resistance, or, to be more accurate, relative resistance at high temperatures, may be similar to that of the phenomenon of rust resistance, the resistance being determined by the formation, or otherwise, and the characteristics of, a protective film of a composition determined, in the first place, by the analysis of the steel, and in the second place, by the composition of the gaseous media. Experimentally, the subject is difficult, if only on one account: great care may be taken in purifying the gases used to produce the atmosphere required, but it should always be remembered that all the steels contain carbon, sulphur and hydrogen in varying, though in the latter cases, small, percentages, and at the high temperatures considered, these elements react with the gaseous medium. Thus small quantities of the products which such elements may form, will be present, and will require to be considered in any theoretical handling of the results. The importance of such percentages of substances, other than those directly under study, has been well emphasized in the investigations of corrosion at ordinary temperatures, of Friend, Evans, and others; it was established by Ariëns Kappers in 1872, that sodium and potassium did not oxidize in dried oxygen, by Cooper in 1882 that chlorine was without action on several metals in the absence of water; while Baker in 1883 established the astonishing fact that purified charcoal could be heated to redness in dried oxygen without burning.

As regards the actual experiments to which the author will now refer, he would say that, in all cases, unless otherwise stated, the samples were approximately 1.25 centimeters in diameter, and weighed about 30 grams (1 ounce); the index figure given is the increase in weight in milligrams per square centimeter of surface. A standard finish, viz., 00 emery, was given to all the cylinders. Thus the data obtained are all strictly comparable. The duration of the exposure was twenty-four hours.

**Table XXXII**  
**Influence of the Atmosphere, Pure and with Various Additions, on a**  
**Mild Carbon Steel (G)**

Condition of Atmosphere	Weight of Samples, Grams	Increase of Weight, Grams	Index Figure
Pure air	21.7344	0.5800	55.238
Atmosphere	21.0978	0.6117	57.166
Pure air + 2% SO <sub>2</sub>	20.8553	0.6843	65.173
Atmosphere + 2% SO <sub>2</sub>	19.7816	0.6905	65.762
Atmosphere + 5% SO <sub>2</sub> + 5% H <sub>2</sub> O	20.4646	1.5242	152.420
Atmosphere + 5% CO <sub>2</sub> + 5% H <sub>2</sub> O	22.4074	1.1048	100.436
Pure air + 5% CO <sub>2</sub>	22.1328	0.9303	76.884
Pure air + 5% H <sub>2</sub> O	20.3436	0.8089	74.211

With a view to determining the effect of the influence of sulphur dioxide, carbon dioxide and H<sub>2</sub>O in facilitating scaling, the author exposed a series of mild steel test pieces and chromium-nickel rustless steel test pieces at 1650 degrees Fahr. (900 degrees Cent.), with the results shown in Tables XXXII and XXXIII. These data do indicate the acceleration of attack brought about by the presence of these influences. It was realized that it was desirable

**Table XXXIII**  
**Influence of the Atmosphere, Pure and with Various Additions, on a**  
**Chromium-nickel Rustless Steel (N)**

Condition of Atmosphere	Weight of Sample, Grams	Increase of Weight, Grams	Index Figure
Pure air	34.7058	0.0062	0.403
Atmosphere	34.5083	0.0071	0.461
Pure air + 2% SO <sub>2</sub>	25.5649	0.0107	0.859
Atmosphere + 2% SO <sub>2</sub>	23.4008	0.0136	1.126
Atmosphere + 5% SO <sub>2</sub> + 5% H <sub>2</sub> O	23.7758	0.0439	3.578
Atmosphere + 5% CO <sub>2</sub> + 5% H <sub>2</sub> O	31.5332	0.0678	4.582
Pure air + 5% CO <sub>2</sub>	31.3954	0.0176	1.175
Pure air + 5% H <sub>2</sub> O	31.1554	0.0516	3.240

to endeavor to couple up the response at ordinary temperatures with that at 1650 degrees Fahr. (900 degrees Cent.); but to do this satisfactorily with the whole of the gaseous media was a very big task, and it was therefore decided simply to experiment, at this stage, with the ordinary atmosphere, but using three steels, i.e., a mild carbon steel, a 14 per cent chromium steel, and the one used in the previous experiment, containing 18 per cent chromium and 8 per cent nickel. The results will be found in Table XXXIV. It is clear from these figures, that the chromium steel (K) is mark-

**Table XXXIV**  
**Atmospheric Tests on Three Steels**

Testing Temperature		Mild Steel	Chromium Steel	Nickel-Chromium
Degrees Fahr.	Degrees Cent.	(G) Index Figure	(K) Index Figure	Steel (N) Index Figure
212	100	nil	nil	nil
392	200	0.033	0.013	nil
572	300	0.127	0.040	0.020
752	400	0.454	0.081	0.040
932	500	0.622	0.093	0.040
1112	600	4.636	0.200	0.133
1292	700	11.917	0.399	0.223
1472	800	44.914	0.767	0.395
1652	900	57.166	1.073	0.461
1832	1000	135.777	66.671	21.820
2012	1100	208.000	165.270	72.301
2192	1200	399.877	261.000	177.660

edly superior to the mild steel (G), but that whilst both steels are unaffected at 212 degrees Fahr. (100 degrees Cent.), they are both attacked at 392 degrees Fahr. (200 degrees Cent.); though in different degrees. There is a great acceleration in the attack on mild steel on approaching 1110 degrees Fahr. (600 degrees Cent.), but the similar acceleration in the case of the chromium steel is postponed until 1658 degrees Fahr. (900 degrees Cent.). A comparison of the results obtained with the nickel-chromium steel (N), indicates clearly the increased resistance obtained by the further modification in composition. The outstanding feature of these experiments, to the author's mind, is the definite attack obtained at such low temperatures in the case of all three steels. According to common conceptions, the zone of temperature from 400 to 930 degrees Fahr. (200 to 500 degrees Cent.) would hardly be associated with the heat resisting problem under consideration, yet clearly the attack appears to be of the same nature as that which takes place at the higher temperatures, but simply of a different order.

As regards the resistance of materials to industrial gases, continuous investigations are in hand in their laboratories, and probably it will be useful to present some of the data. One of the conditions of test is the atmosphere of an unenclosed gas muffle, and the composition which may be taken as representative of the atmosphere, is as follows.

	Per Cent
Nitrogen .....	67.70
Oxygen .....	1.34
Carbon dioxide .....	4.75
Steam .....	21.10
Sulphur dioxide .....	0.003
Carbon monoxide .....	5.10

Table XXXV  
Analyses of Materials

Sym- bol	Material	Condition	C	Mn	Si	S	P	Per Cent Cr	Ni	W	Cu	Fe
A	Pure iron	Normalized 1742 degrees Fahr.	0.04	0.01	0.01	nil	nil	.....	0.02	.....	.....	99.92
B	Chromium	As cast	0.16	0.02	0.40	.....	.....	98.10	.....	.....	.....	1.30
C	Nickel	As cast	0.04	.....	.....	.....	.....	.....	99.58	.....	.....	0.38
F	Tungsten	Hard drawn	.....	nil	.....	.....	.....	.....	.....	99.99	.....	.....
G	Mild carbon steel	Normalized 1652 degrees Fahr.	0.17	0.67	0.18	0.012	0.018	.....	0.25	.....	.....	98.70
H	3% nickel steel	Oil hardened 1562 degrees Fahr. tempered 1166 degrees Fahr.	0.34	0.60	0.29	0.021	0.012	.....	3.25	.....	.....	95.49
I	36% nickel steel	Air cooled 1742 degrees Fahr.	0.24	0.40	0.16	0.020	0.015	0.09	36.90	.....	.....	62.17
J	Silicon-chromium steel	Air cooled 1922 degrees Fahr. tempered 1562 degrees Fahr.	0.50	0.37	3.04	0.01	0.01	8.28	0.16	.....	.....	87.63
K	Chromium steel	Oil hardened 1832 degrees Fahr. tempered 1472 degrees Fahr.	0.32	0.25	1.32	0.009	0.016	13.12	0.29	.....	.....	85.67
L	Chromium steel	Air cooled 1652 degrees Fahr.	0.09	0.39	0.37	0.011	0.015	18.53	0.26	.....	.....	80.334
M	Chromium-nickel steel	Air cooled 2102 degrees Fahr.	0.11	0.34	0.21	0.011	0.013	14.84	10.16	.....	.....	74.816
N	Chromium-nickel steel	Air cooled 2102 degrees Fahr.	0.12	0.28	0.31	0.008	0.014	17.74	8.06	.....	.....	73.47
O	Nickel-chromium steel	As rolled	0.35	1.36	0.21	0.021	0.013	10.90	35.15	.....	.....	52.00
P	Chromium-nickel-silicon steel	Water quenched 1832 degrees Fahr.	0.58	0.36	4.00	0.020	0.017	15.54	8.14	.....	.....	71.43
Q	Chromium-nickel-tungsten steel	As rolled	0.30	0.52	1.46	0.02	0.017	17.74	7.0	4.23	.....	68.713
R	Nichrome	Air cooled 1742 degrees Fahr.	0.03	0.89	0.71	.....	.....	11.69	60.40	.....	.....	26.24
S	Cast iron	As cast	2.52	0.72	1.13	0.125	0.58	.....	.....	.....	.....	94.24
T	15% silicon iron	Gr. C. C. C. Gr. C. C. C.	0.68 0.76 0.76 0.03	0.21	14.30	0.045	0.731	.....	.....	.....	.....	83.92
U	Monel metal	Hot rolled	0.14	1.00	0.63	.....	.....	.....	69.98	.....	27.17	1.67

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**Table XXXVI**  
**The Influence of Industrial Gases**

Symbol	Material	Complex gas at 1652° F. Index figure	Muffle at 1652° F. Index figure
A	Pure iron	114.012	95.34
B	Chromium	2.051	1.00
C	Nickel	5.222	16.83
G	Mild carbon steel	80.225	73.104
H	3% Nickel steel	72.893	43.30
I	36% Nickel steel	30.058	27.70
J	Silicon-chromium steel	0.84	0.506
K	Chromium steel	18.319	20.50
L	Chromium steel	1.252	1.38
M	Chromium-nickel steel	6.000	3.84
N	Chromium-nickel steel	2.699	0.326
O	Nickel-chromium steel	2.534	1.01
P	Chromium-nickel-silicon steel	0.689	0.061
Q	Chromium-nickel-tungsten steel	0.43	0.175
R	Nichrome	1.397	0.765
S	Cast iron	86.000	33.80
T	15% Silicon iron	78.434	13.734
U	Monel metal	68.676	1.55

The other atmosphere is that produced by a complex gas, synthetically produced as representing the products of combustion, of the following composition:—

	Per Cent
Nitrogen .....	72.95
Oxygen .....	5.00
Carbon dioxide .....	12.00
Steam .....	10.00
Sulphur dioxide .....	0.05

In Table XXXV will be found a statement concerning the analyses and conditions of the materials tested, and it will be seen that the list has been so put together as to make easy reference possible with regard to the resistance to scaling of the materials generally available.

In Table XXXVI will be found a statement of the actual index figures obtained from this series of results in the two atmospheres, at 1652 degrees Fahr. (900 degrees Cent.). On considering the influence of these industrial gases on the various samples, it will be seen that, although only one temperature, 900 degrees Cent., is being considered, there are interesting facts to be discerned. Pure iron (A), as would be expected, is very badly attacked, chromium (B) is very resistant, while compared with

iron, nickel (C) gives a good account of itself. Mild carbon steel (G) is more resistant than the pure iron (A), possibly on account of the protection afforded by its carbon content. The influence of nickel is well illustrated by the steels containing 3.0 and 36 per cent respectively. (H and I), and it will be seen that the effect is undoubtedly to produce increased resistance. The silicon-chromium steel (J) is, curiously, much more resistant even than chromium. In the two chromium steels (K and L), 13 per cent chromium undoubtedly had a profound influence upon the behavior of the alloy, but by increasing the percentage to 18.0, a resistance of the same order as that of chromium (B) is obtained. On studying the steels (M), (N) and (O), containing both nickel and chromium, a high but variable resistance due to the variable proportions of the elements is found; (N) gives the most satisfactory results, but it is interesting to compare (M) and (K), in which the chromium is roughly comparable, but in which there is the outstanding variable of 10 per cent nickel in (M), and a greatly improved resistance. The optimum results obtained from (N) are seen to be still further improved in the cases of either (P) or (Q) by the addition of silicon and tungsten respectively. Nichrome, as would be expected, gives very good results, but it is not equal to either (J), (P) or (Q), which are charged with a far smaller amount of the added metals. The cast iron (S), the 15 per cent silicon iron (T), and the Monel metal results are of interest by way of comparison with the general series.

It is to be deduced from these figures that the resistance of iron is greatly increased by the addition of either chromium or nickel, or both; the addition of silicon or tungsten still further increases the resistance of such alloys.

A comparison of these results would indicate quite clearly that the "complex" gas, as a general representative of products of combustion, is much more active in attacking the samples than the muffle atmosphere: the latter series of tests was introduced to enable a comparison with the other more "experimental" tests to be made in an empirical practical manner. Broadly speaking, the greater activity of the "complex" gas must be accepted, and it is, therefore, of interest to compare the two gaseous atmospheres. The "complex" gas contains more carbon dioxide, oxygen, and sulphur dioxide, and very much less steam. The author is of the

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Chemistry

opinion that the increased attack is due to the increased percentages of  $\text{CO}_2$  and  $\text{SO}_2$ , particularly of the latter.

In one of the author's papers to which he has just referred,<sup>161</sup> an account is given of fundamental experiments with a view to deducing the effect of the various elements. The alloying of chromium with iron in the production of rustless and heat resisting steels, has been previously well dealt with in technical literature. The patent history of this and other countries also yields much information. The work of Brearley, Haynes, Pasel, Becket, the author, and others, laid a sound foundation to the knowledge of that particular field. An interesting paper by McQuigg<sup>162</sup> discusses the influence of chromium in increasing the heat-resistance of steels, while the American Society for Testing Materials Symposium of 1924 did much to focus attention on the subject, notable papers being given by Johnson and Christiansen, Hunter and Jones, and Fahrenwald. The American Symposium, to which reference has just been made, has also done much to extend the use of alloys containing nickel in addition to chromium, thus encouraging the utilization of the additive advantage of that element. The apparently anomalous effect of nickel in assisting in the protection of the steel of which it is a constituent, may probably be explained by the influence of that element in causing the formation of a particular kind of protective layer. That nickel is a valuable element in these heat resisting steels has also been established by other investigators. Sir Robert Hadfield,<sup>163</sup> went so far as to suggest that the inclusion of nickel was responsible for a great step forward. Monypenny<sup>164</sup> described the increased resistance due to the addition of this element; and the same deduction is also to be made from the paper by Elliott and Willey.<sup>165</sup>

The influence of silicon and tungsten in further increasing the resistance, would appear to be fully established. In the case of tungsten, this effect is certainly anomalous, since, in the author's fundamental experiments, an abnormal attack in oxygen was dis-

<sup>161</sup>W. H. Hatfield, *Journal*, Iron and Steel Institute, Vol. CXV, 1927.

<sup>162</sup>McQuigg, *Proceedings*, American Institute of Mining and Metallurgical Engineers, Vol. LXIX, 1923, p. 831.

<sup>163</sup>Robert Hadfield, Paper read before the Institute of Marine Engineers, January 11, 1927.

<sup>164</sup>Monypenny, "Stainless Iron and Steel," London, 1926, Chapman and Hall, Ltd.

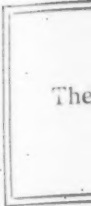
<sup>165</sup>Elliott and Willey, Paper read before the Congress of Chemists, July 20, 1926: *Chemistry and Industry*, July 30, 1926, p. 526.

closed, a heavy attack in steam, but a materially diminished attack in  $\text{CO}_2$  and  $\text{SO}_2$ . It might, of course, be deduced that the additive protection accorded by the metal is in line with its behavior toward the anhydrous acids  $\text{CO}_2$  and  $\text{SO}_2$ , particularly the latter.

The main feature of these notes is the attempt to establish quantitatively the fact that these added elements do indeed produce steels which satisfactorily resist attack at high temperatures, and among such can safely be cited certain compositions in the chromium, the silicon-chromium, chromium-nickel, chromium-nickel-silicon and chromium-nickel-tungsten series. Some of these steels have extremely satisfactory characteristics as regards resistance to corroding influences under the conditions studied, and are fitted for much wider service than that in which they are at present employed.

In a consideration of this work, it should be borne in mind that, apart from the simple metals, the materials selected for the research largely represent actual steels which have seen service conditions. It will be realized that with so many elements to work upon, the combinations are almost without limit. It is the author's opinion that a proper appreciation of the possibilities of this field will enable the engineering world to develop successfully the various projects so dependent upon the satisfactory resistance of steel at high temperatures. Some of the steels dealt with in the paper are already complex in composition and constitution, but experiment may finally establish that the addition of cobalt, molybdenum, aluminum, and other elements may result in further advantages as regards the desirable characteristics. Suffice it to have shown that the last few years have produced some special steels largely fulfilling the requirements of high strength at high temperatures, coupled with an excellent degree of resistance to scaling under such conditions.

(To be concluded in August issue of TRANSACTIONS)



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## Educational Section

These Articles Have Been Selected Primarily For Their Educational  
And Informational Character As Distinguished From  
Reports Of Investigations And Research

### THE CONSTITUTION OF STEEL AND CAST IRON SECTION II—PART X

By F. T. SISCO

#### *Abstract*

*This installment, the tenth of the present series, discusses the operation of carburizing. The first question of importance is that of the composition of the steel. The low carbon steels are used for carburizing because they give a hard case and a soft ductile core. The next question concerns the chemical reactions taking place when steel is carburized with solid cements, and this is followed by a discussion of the various compounds commonly used. The installment closes with a discussion of carburizing temperatures and the depth and chemical composition of the case.*

WHEN a piece of high carbon steel is quenched in water from above the critical range, it becomes very hard but at the same time it becomes very brittle. It is, of course, possible to reduce the brittleness by tempering and the higher we temper this quenched steel the more ductile it becomes; but as the ductility increases the hardness and strength decrease proportionally. For many purposes where a high hardness is required, as in some classes of tools, a certain amount of brittleness does no harm and so is tolerated, but there are also many uses to which steel is put where a high hardness and great toughness are both necessary. In general it is not possible to secure the contradictory properties of high hardness and great toughness by our standard methods of hardening and tempering,

This is the twentieth installment of this series of articles by F. T. Sisco. The several installments which have already appeared in TRANSACTIONS are as follows: June, July, August, September, November, 1926; January, February, April, June, August, October, 1927; February, April, June, November, 1928; January, March, May, June, 1929.

The author, F. T. Sisco, a director of the society, is chief of metallurgical laboratory, Air Corps, War Department, Wright Field, Dayton, Ohio.

hence recourse must be had to a special method of securing this combination of properties. This special method is known as carburizing, sometimes as cementation.

The process of carburizing depends upon the fact that when a steel of proper composition is heated in close contact with a suitable material containing carbon, the carbon will pass into the steel where it unites with the iron according to the reaction



forming iron carbide. The amount of carbon that passes into the piece being carburized and the distance it penetrates in from the surface depends upon the nature of the carburizing compound, the temperature to which the material is heated and the length of time it is held at that temperature.

The result of this carburization is that at and near the surface, the steel is high in carbon, often containing as much as 1.00 per cent. Beginning just in from the surface is a zone in which the carbon graduates from high to low. The interior portion, which is usually a low carbon steel, is unaffected. Now if this carburized material is heat treated by quenching from a temperature of 1400 to 1500 degrees Fahr. followed by tempering just high enough to relieve strains we have as a result a section which has a very hard, wear-resisting surface and a tough ductile core. Carburization while extensively practiced is more costly than a simple hardening and tempering operation, consequently its use is restricted to parts subjected to hard wear and great shock such as gears, pinions, piston pins, splines and the like.

The process of carburizing consists of packing the machined steel parts with some carbonaceous material known as the carburizing compound in a box or suitable container, sealing the box to prevent ingress of air and heating in a furnace to the proper temperature and for the proper time. After the steel is carburized to the required depth, the part is subjected to heat treatment in order to develop the most desirable combination of properties in both case and core. It is needless to state that steel articles to be carburized must be machined either before carburizing or before the hardening operation. The only shaping possible after heat treatment is by grinding.

In discussing carburization, it is necessary to take into account

four factors (1) the composition of the steel (2) the carburizing compound (3) the carburizing temperature, and (4) the length of time that the material is held at the required temperature. The heat treatment which follows naturally depends upon the use to which the steel is to be put. There are two general divisions of case carburization. The first and most common is carburizing with charcoal or some allied compound; the second is carburizing with cyanide.

#### COMPOSITION OF THE STEEL

As one of the desiderata of a carburized steel article is ductility or toughness, it is essential that we start with a steel which is inherently soft and ductile, consequently the carburizing steels are generally low in carbon. The plain carbon steels containing between 0.08 and 0.25 per cent carbon are most common. Next to these come the low carbon 3.50 per cent nickel steels, the low carbon chromium-nickel and chromium-vanadium steels.

From the standpoint of carbon penetration there is no reason why we cannot use a steel that contains more than 0.25 per cent carbon. The chief objection to the medium and high carbon steels is that they respond to heat treatment with the result that the core becomes too hard and brittle. For this reason the low carbon steels are invariably used. The dead soft steels, 0.10 per cent carbon or below are undesirable because they are difficult to machine.

One notable exception to the general rule of using low carbon steels for carburizing is found in armor plate for battleships. The steel used a few years ago for this purpose contained 0.32 to 0.40 per cent carbon, 3.50 to 4.00 per cent nickel, and 1.75 to 2.25 per cent chromium. By the proper methods plates made from this steel may be carburized as deeply as one to 1½ inches. Armor plate presents a special problem in carburizing and the practice is constantly changing. This is due to the endless race between armor-piercing projectiles and the hardened plate. As the projectiles increase in penetrating power, the hardness and resistance of the plate must be increased to meet it.

In general alloy steels carburize well. Nickel increases the strength and toughness of the core. Chromium tends to produce a harder case but sometimes makes the material difficult to machine before heat treatment. Steels of the stainless variety, even the stainless irons practically free from carbon, are difficult to carburize

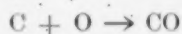
satisfactorily. It is thought that manganese makes the case of a carburized steel brittle, consequently the amount of this element is usually maintained well below 0.60 per cent. Sulphur and phosphorus should be low in steels to be carburized. Banding and segregated ferrite are commonly encountered in steels containing more than 0.050 per cent phosphorus. When these structural defects are present the carbon penetration frequently is not uniform.

It is generally accepted that satisfactory carburizing does not result from the use of steel containing oxides, gas, slag and other impurities and inclusions. Ehn<sup>56</sup> maintains that oxide is responsible for abnormal carburizing properties and the soft spots that result from heat treating these abnormal steels. We shall have occasion to look at these steels in detail in just a few moments.

#### THE CARBURIZING REACTION

If we pack a piece of low carbon steel in a box in contact with charcoal or other pure carbon and evacuate the box so that there is no air or other gas present and heat it to the carburizing temperature, a small amount of carbon will eventually pass into the steel. While it is theoretically possible to carburize to a slight degree with pure carbon, it is economically impracticable because the carburization taking place under these conditions is negligible. The reason for this is apparent when we consider the nature of the steel. Carbon as such migrates into steel with difficulty, but when heated the steel becomes easily permeable to gas, which readily diffuses into the metal. Thus in commercial carburization we take advantage of this and use gas as the carrier for the carbon.

Let us consider the simplest possible case, carburizing with ordinary wood charcoal. The steel to be carburized is packed in a box and completely surrounded with the charcoal. The box is then sealed and heated to the proper temperature. What happens? The oxygen in the air occluded in the box during the packing operation will react with the carbon in the charcoal at a bright red heat to form carbon monoxide, CO,



This carbon monoxide will diffuse into the steel where it reacts with the iron according to the reaction,

<sup>56</sup>E. W. Ehn, "Irregularities in Case Hardened Work Caused by Improperly Made Steel," *TRANSACTIONS, American Society for Steel Treating*, Vol. 2, 1922, p. 1177-1202. See also *Transactions, American Institute of Mining and Metallurgical Engineers*, 1922, p. 341; and *Journal, British Iron and Steel Institute*, May, 1922, p. 157.



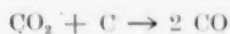


or we can put the reaction in this way,



In this latter reaction we might say that iron has a catalytic action upon the carbon monoxide and sets free one atom of carbon.

The carbon dioxide,  $\text{CO}_2$  which results from the above two reactions reacts with more carbon in the charcoal.



Thus as long as we have sufficient charcoal present and enough oxygen (air) present in the box to start the reactions and as long as the temperature is high enough the process will continue almost indefinitely. Thus it is plain that solid carbon does not carburize a piece of steel; but the carbon monoxide gas resulting from the reaction between solid carbon and oxygen does carburize the steel.

In the air occluded in the box there is nitrogen as well as oxygen. Under certain conditions and with steels of certain composition, the metal will absorb nitrogen but at the temperatures usually employed for carburizing (1600 to 1850 degrees Fahr.) this gas has little or no effect. Some investigators claim that nitrogen exercises a deleterious effect, probably by the formation of a brittle iron nitride. For practical purposes, however, it is safe to assume that in ordinary carburizing, nitrogen has no action at all.

#### CARBURIZING COMPOUNDS

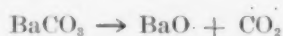
Since carburizing depends upon a gas it is natural to use some material which generates its own gas. For this purpose a large number of carburizing compounds have been made. Some of these are very effective, others moderately so, while some for which extravagant claims are made are palpable nostrums. One of the most effective and satisfactory carburizing compounds is made of wood charcoal and barium carbonate. Among the other substances which may be added to charcoal are salt, potassium nitrate, cyanide and ferrocyanide. These various chemical compounds are known as energizers or activators. The function of the energizer is not clear in every case; it is certain however that it hastens carbon penetration.

Concerning the forms of carbon used, wood charcoal is the most common but many prefer the charcoal resulting from organic mate-

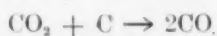
rial such as charred leather, bone charcoal, animal black, charred sugar and the like. These charcoals are usually considered to be more effective than the product of wood.

Carburizing compounds are usually purchased all prepared and ready for use. Commonly they are in the form of small oval nodules about the size of a pea. Each particle of charcoal is impregnated with the proper amount of energizer. In commercial case carburization, the best results are obtained by carburizing with a compound made of wood charcoal and barium carbonate, using for each heat 20 to 50 per cent new compound mixed with the used material.

Although the action of many of the energizers is not known exactly, in the case of barium carbonate, the most common one, the mechanism has been thoroughly studied. Under the action of heat the barium carbonate dissociates:



The carbon dioxide,  $\text{CO}_2$ , formed reacts with the carbon of the charcoal



The carbon monoxide gas,  $\text{CO}$ , penetrates the steel where it gives up part of its carbon.



The carbon dioxide reacts with the carbon of the charcoal,



and thus the process continues.

Since the carburizing action is due to the penetration of the steel by a carbonaceous gas, it is evident that we can carburize steel by the use of a gas alone, without the presence of solid carbon. This method is also used for commercial carburization. When carbon monoxide gas alone is used for carburization, the penetration is even but the carbon content of the case is relatively low, usually below the eutectoid point. The carbon content of the case and the rate of penetration are directly dependent upon the temperature and the gas pressure.

One of the most effective carburizing agents is gas composed of the hydrocarbons, methane and allied compounds of this series. Natural gas is commonly used. The hydrocarbons often produce a high carbon case, usually well above the eutectoid point, and some-

times a thin transition layer which contains approximately the eutectoid ratio. Carburizing with natural gas consisting principally of hydrocarbons has the disadvantage that if carried on at a low temperature, just above the critical range for example, the process is slow and the case is not uniform; if carried on at high temperatures the case is too high in carbon. Most of the animal charcoals such as charred leather, bones, horns and the like give off complex hydrocarbons when heated.

In discussing the various compounds Bullens says<sup>57</sup> that one of the best solid cements for general use is composed of 10 to 30 per cent barium carbonate and the balance hardwood charcoal. A small amount of calcium and sodium carbonate is frequently added. This compound produces a carburized zone which is markedly more homogeneous than the other solid cements. The carbon contents of the case will vary according to the temperature of carburizing. Bullens gives a minimum carbon of 0.70 per cent for very thin zones obtained with a temperature of 1650 degrees Fahr. to a maximum of 1.3 per cent for zones thicker than 0.04 inches obtained with a temperature of 2000 degrees Fahr. The wood charcoal, barium carbonate compound has the advantage that it regenerates itself when exposed to the air. The barium oxide which is left after the dissociation of the barium carbonate absorbs carbon dioxide from the air thus returning to its original condition as barium carbonate. Most of the carburizing compounds on the market are wood charcoal-barium carbonate type.

#### CARBURIZING TEMPERATURES, CASE DEPTH AND COMPOSITION

In commercial carburizing, the case depth varies from 1/64 inch to 1/16 inch or even more in exceptional cases. The penetration depends upon the temperature and the time of exposure. In general the higher the temperature and the longer the time that the material is held at this temperature the deeper the case. While this holds true in a general way it should be noted that the depth of penetration does not always increase in direct proportion to the time and temperature. For example consider the following data collected by Giolitti using a wood charcoal, barium carbonate compound.

<sup>57</sup>D. K. Bullens, "Steel and Its Heat Treatment," 1918, p. 235.

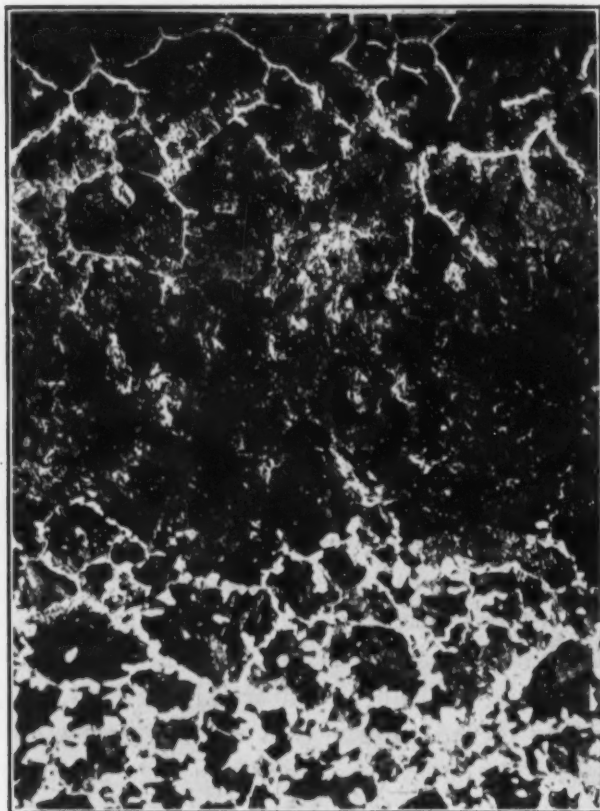


Fig. 70—Typical Structure of a Carburized Low Carbon Steel, Slowly cooled.  $\times 100$ . Etched with Alcoholic Nitric Acid.

Hours	Depth of case at temperature indicated (In Inches)		
	1550 °F.	1650 °F.	1800 °F.
1	0.020	0.030	0.045
2	0.035	0.044	0.060
3	0.044	0.052	0.075
4	0.050	0.062	0.088
5	0.052	0.068	0.100
6	0.056	0.072	0.110
7	0.060	0.078	0.120
8	0.062	0.082	0.130

When carburized at 1550 degrees Fahr. the penetration is nearly proportional for two to three hours but longer times produce very little additional case. When carburized at 1650 degrees Fahr. the penetration is nearly proportional for three to four hours. Holding for a longer time produces a somewhat greater penetration but at a decreased rate. At 1800 degrees Fahr. however, there is rapid penetration for the first four hours and upon further heating



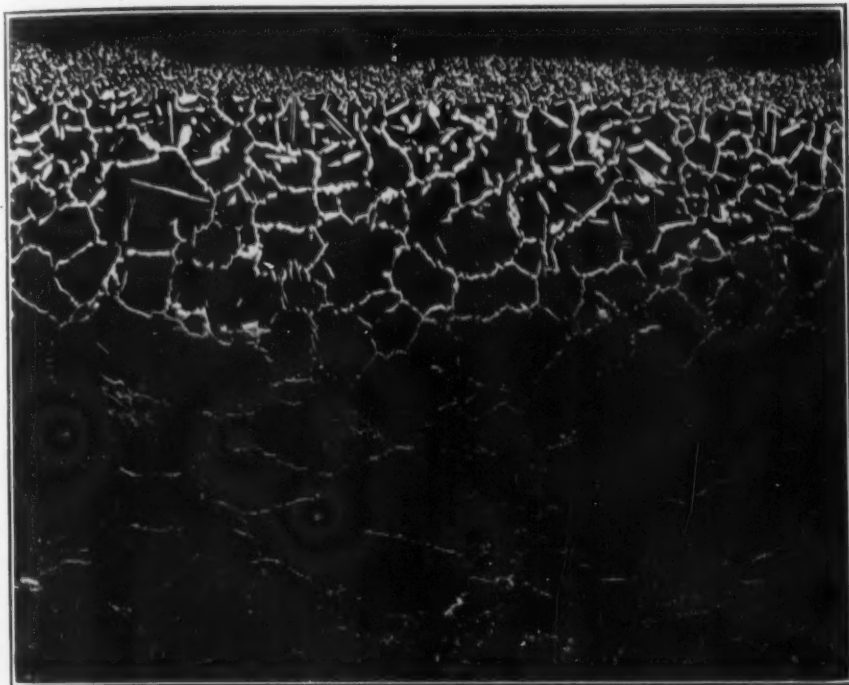


Fig. 71—Typical Structure of a Carburized Medium Carbon Chromium-nickel Steel Armor Plate, Slowly Cooled.  $\times 100$ . Etched with Alcoholic Nitric Acid.

a constant and but slightly less marked penetration takes place. It should be remembered that the figures shown in the table are accurate only for the particular compound used. The effect of the compound upon the case depth has been fully investigated. The following taken from Lake is perhaps typical of the effect of the carburizing compound upon the case depth.

Hours	Rate of Penetration at 1650 Degrees Fahr. (In Inches)		
	Charred Leather	Wood Charcoal	Wood Charcoal and $\text{BaCO}_3$
2	0.045	0.028	0.055
4	0.062	0.042	0.087
8	0.080	0.062	0.111
12	0.110	0.070	0.125

The American Society for Steel Treating has worked out a table for normal carbon carburizing steels.<sup>58</sup> This table gives the results obtained in ordinary practice with commercial compounds and shows the case depth resulting from various temperatures. As representative of these results the following temperatures and times give a case of  $\frac{1}{16}$  inch.

<sup>58</sup>American Society for Steel Treating *Handbook*, 1929 Edition, p. 204.

Degrees Fahr.	Hours
1600	13
1650	10
1700	8
1750	6
1800	5
1850	4

The customary temperature range for carburizing is 1600 to 1750 degrees Fahr. and the average carbon penetration at these temperatures is 0.008 to 0.015 inches per hour.

It is evident from these results that the depth of case depends upon the temperature and the time. In general the amount of carbon taken up by the steel is dependent upon the carburizing compound used. This factor also affects the carbon gradient from the outer edge of the case to the low carbon zone of the original steel. These factors have been worked out carefully for the commercial carburizing compounds. It is desirable not to have a too high carbon content in the case, in general it should not exceed 1.10 per cent, and even a lower percentage than this may be desirable. It is also desirable to avoid a too abrupt change in the distribution of the carbon from the hypereutectoid zone near the surface to the hypoeutectoid zone in the center. When this change is abrupt we may have a plane of weakness between the two where the carbon content is approximately 0.90 per cent.

A typical structure of a slowly cooled carburized low carbon steel is shown in Fig. 70. Near the edge of the specimen (marked by the line) the free cementite (white) has formed a network around the pearlite grains. The transition zone of granular pearlite and sorbite is relatively wide and graduates into the low carbon core which is characterized by the numerous particles of free ferrite. A typical structure of a carburized medium carbon nickel-chromium steel is shown in Fig. 71. This specimen was taken at the edge of a piece of armor plate carburized to a depth of nearly two inches. The specimen had not been heat treated. At the extreme edge the carbon is very high as is evident from the large amount of free cementite. Near the edge the cementite forms a distinct network around the grains of granular pearlite or sorbite, and further from the edge of the specimen there are still traces of free cementite assuming the form of a faint network around the sorbite grains.

## Recommended Practice Committee Releases

### CHROMIUM AS AN ALLOYING ELEMENT IN STEEL

By Marcus A. Grossmann\*

**Metallic Chromium**—Pure chromium is a hard silver-gray metal. The physical constants for the element chromium are given in the data sheet on the Physical Constants of the Principal Alloy Forming Elements.

**Chromium in Steel**—Chromium dissolves in liquid iron, and according to Sauveur is retained in solid solution both in gamma and alpha iron. In the presence of carbon a chromium carbide is formed and when other alloying elements are present double and complex carbides are formed.

The constitutional diagram for chromium steels is given in Fig. 1.

Chromium imparts to steel hardness, wear resisting properties, useful magnetic properties, permits a deeper penetration of hardness, raises the critical points, increases resistance to corrosion, and increases the mechanical property values.

**Source of Chromium**—The principal source of chromium is the ore chromite ( $\text{FeO} \cdot \text{Cr}_2 \text{O}_3$ ), consisting of chromic oxide, with certain proportions of iron oxide and manganese oxide. The ores are widely distributed over the face of the earth and there is an abundance of the material available.

For addition to steel, the ore is converted to ferrochromium which usually contains about 60 per cent chromium and about 30 per cent iron. The balance is mainly carbon and silicon. Ferrochromium is added to the molten steel, in the furnace, in the proper proportion to give the desired alloy content.

The ferrochromiums used in steel making are classified according to their carbon content, which may range from 6.0 down to 0.1 per cent carbon. The basis of classification lies not in variation in the quality of the different grades of ferrochromium but in their cost. It is more difficult to manufacture ferrochromium with a low carbon content. Hence, the ferrochromiums with higher carbon are cheaper, and they are therefore used wherever possible.

**Table I**  
**Chromium Steels—S.A.E.**

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus Max.	Sulphur Max.	Chromium Range
5120	0.15-.25	30-60	.04	.045	.60-.90
5140	.35-.45	50-80	.04	.045	.80-1.10
5150	.45-.55	50-80	.04	.045	.80-1.10
52100	.95-1.10	20-50	.03	.030	1.20-1.50

\*Research metallurgist, Central Alloy Steel Corp., Massillon, Ohio.

The grade of ferrochromium used is dependent on the amount of chromium which must be introduced into the steel and the permissible rise in carbon content of the steel which may accompany it.

**Chromium Automotive Steels**—The straight chromium steels used for automotive purposes are covered by the S. A. E. 5100 series and are given in Table 1. This series together with the 3100, 3200, 3300, 3400, 4100 and 6100 S. A. E. steels includes the various alloy steels containing chromium for carburizing, case hardening, and for use in the heat treated condition.

The following two steels are typical of the low chromium steels of the 5100 series which are used in the heat treated condition.

	C	Cr	Mn	Si
I .....	.30	.90	.70	.25
II .....	.50	1.00	.75	.25

Steel I is generally used in a toughened condition, that is, after quenching it is often tempered from 900 to 1100 degrees Fahr., to a Brinell hardness around 325. Steel II is generally used in a harder condition, the quench being followed by tempering at only 350 to 600 degrees Fahr. with a resulting Brinell hardness often around 525.

The following properties are typical when these steels, in a one inch section, are quenched and tempered as shown:

	Tempering Temp. Degrees Fahr.	Tensile Strength Lbs. per sq. in.	Yield Point Lbs. per sq. in.	Elong. Per Cent in 2 in.	Brinell Hardness
Steel No. I	600	215,000	195,000	13	430
Quenched	800	190,000	175,000	14	390
1525° F.	1000	170,000	150,000	18	325
Water	1200	135,000	110,000	25	250
Steel No. II	600	260,000	235,000	11	512
Quenched	800	235,000	210,000	13	455
1525° F.	1000	190,000	165,000	17	375
Oil	1200	150,000	130,000	22	290

**Chromium-nickel Steels**—The oldest application of chromium was in the chromium-nickel steels for armor plate and for many years these were the only alloy steels widely used. The chromium-nickel steels have come to be used even more widely in the well known automotive steels, and are now used for all kinds of construction. When combined, chromium and nickel are used for the most part in three combinations as follows:

3.50 per cent nickel with 1.50 per cent chromium.
3.50 per cent nickel with 0.75 per cent chromium.
1.50 per cent nickel with 0.75 per cent chromium.

These correspond to the classification of the Society of Automotive Engineers known as S. A. E. 3300, 3400 and 3100. The carbon content of the steels may range from 0.15 to 0.55 per cent from the class of case hardening to that of oil-hardening. Another steel, used widely for case hardening and known as Krupp analysis, contains 4.0 per cent nickel, 1.5 per cent chromium and 0.08 to 0.15 per cent carbon.

The steels with the lesser amounts of alloy and the lower carbon contents are in general water hardening,—that is, the best treatment involves quenching into water from the hardening temperatures. The steels with high alloy



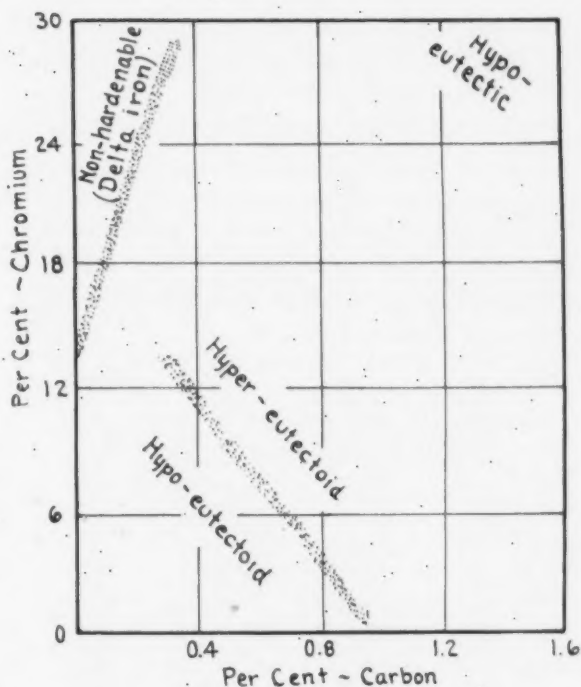


Fig. 1—Constitutional Diagram for Chromium Steels.

contents and with the high carbon are hardened by quenching into oil. Hardening temperatures vary from 1500 to 1650 degrees Fahr., depending on the composition and on the size.

The following three steels may be considered typical of the chromium-nickel series.

S.A.E. No.	C	Cr	Ni
3330 .....	0.30	1.50	3.50
3130 .....	0.30	0.60	1.25
3140 .....	0.40	0.60	1.25

With these steels, the following properties are typical when the steels, in a one inch section, are quenched and tempered as shown:

	Tempering Temp. Degrees Fahr.	Tensile Strength Lbs. per sq. in.	Yield Point Lbs. per sq. in.	Elong. Per Cent in 2 in.	Brinell Hardness
Steel 3330	600	205,000	174,000	13	370
Quenched	800	181,000	157,000	15	335
1500° F.	1000	142,000	123,000	20	270
Oil	1200	115,000	99,000	24	220
Steel 3130	600	212,000	190,000	14	380
Quenched	800	182,000	165,000	17	340
1500° F.	1000	150,000	150,000	21	280
Water	1200	125,000	105,000	25	225
Steel 3140	600	220,000	185,000	9	390
Quenched	800	180,000	148,000	12	330
1475° F.	1000	130,000	105,000	17	260
Oil	1200	110,000	90,000	22	220

**Chromium-Vanadium, Chromium-Molybdenum, Chromium-Silico-Manganese Steels**—The chromium-vanadium and chromium-molybdenum steels

are used for purposes similar to those of the chromium-nickel steels. The former finds further application as spring steels, and for this work chromium-silico-manganese steels are also used widely.

For showing tensile properties, the following steels are typical of this group:

	C	Mn	Cr	V	Mo
I .....	0.38	0.68	1.02	0.17	....
II .....	0.50	0.75	1.00	0.15	....
III .....	0.50	0.75	1.00	....	0.25

The following properties are typical when these steels, in a one inch section, are quenched and tempered as shown:

Steel	Tempering Temp. Degrees Fahr.	Tensile Strength Lbs. per sq. in.	Elastic Limit Lbs. per sq. in.	Elong. Per Cent in 2 in.	Brinell Hardness
I Quenched 1525-1575, oil	600	245,000	203,000	11	444
	800	203,000	187,000	14	395
	1000	180,000	165,000	17	341
	1200	130,000	122,000	20	263
II Quenched 1600, oil	600	250,000	210,000	8	500
	700	239,000	202,000	..	485
	800	220,000	190,000	10	470
	900	200,000	185,000	..	440
III Quenched 1650, oil	1000	180,000	160,000	15	410
	600	260,000	230,000	10	485
	800	235,000	215,000	11	455
	1000	205,000	190,000	14	415
	1200	170,000	155,000	19	345

**Ball Bearing Steels**—The prime requisite for ball bearing steels is high hardness. To impart this property, chromium is used with high carbon without other alloys. The steel must also have adequate crushing strength, but there is generally little difficulty in this respect if the steel is sound and is well made. The ball steels and the bearing steels are quite similar in composition, though the ball steels are often lower in chromium and in carbon than the bearing steels.

The following compositions are typical of these steels:

	C	Mn	Si	Cr
Ball Steel .....	0.90-1.05	0.30-0.45	0.10-0.25	0.75-1.25
Bearing Steel ....	1.00-1.15	0.30-0.45	0.10-0.25	1.25-1.60

Roller bearing steels may have compositions quite similar to these, but may also be much lower in chromium and in carbon.

The steels are hardened by heating to 1450 degrees Fahr. for small sections, up to 1550 degrees Fahr. for larger sections, and then quenching from these temperatures into light oil. The lower chromium steels may be quenched into cold water. They are tempered at low temperatures, in the neighborhood of 300 degrees Fahr.

The outstanding feature of these steels is their great hardness. When properly heat treated they show a Rockwell-C hardness of 65 to 68 and a scleroscope hardness of 100 to 110. Ordinary Brinell readings give fictitious values, because the Brinell ball itself is made of this type of steel.

**Stainless Steels and Alloys**—When chromium is used in adequate percentages, it imparts to steels great resistance to corrosion. In many stain-

less steels and alloys, chromium is used without other alloy, and it at least constitutes one of the alloying elements in all such materials in commercial use today.

Stainless steel found its first wide application in cutlery, where the advantages of cleanliness immediately established the steel in that field. It is now coming into use where other problems of corrosion must be met, in such widely diversified applications as pumps for acid mine waters, piston rods for steam engines, exhaust valves for gasoline engines, instruments for surgical and dental work, and certain classes of decorative hardware and furnishings.

The compositions in commercial use today cover a wide range, but in general may be included under the following types:

#### Stainless Steels and Alloys

C	Cr	Si	Ni
0.35	14.00	....	.....
0.10	13.00	....	.....
0.10	17.00	0.8	.....
0.12	18.00	....	8.0
0.35	25.00	....	.....
0.50	12.00	3.00	.....
0.60	7.00	....	20.00
1.00	9.00	4.00	.....
0.40	7.00	1.00	23.00

The steels with the lower carbon contents exhibit the best stainless qualities, but they do not have as great hardness as do those with higher carbon. The steels in the best stainless range contain about 0.35 per cent carbon, and when hardened have a Brinell hardness of about 500.

As the carbon content is increased, the steel corrodes a little more readily, and the stainless qualities are sacrificed somewhat for hardness. When the carbon content is raised to about 1 per cent, it becomes possible to obtain Brinells over 600, Rockwell-C over 65 and scleroscope over 80.

The stainless alloys low in carbon which are not hardened find wide application in sheets, and in drawn and stamped articles. The sheets are annealed to be as soft as possible, for mechanical strength does not come into consideration, whereas high ductility is usually desirable.

For further information see the data sheet on stainless steel in A. S. S. T. Handbook.

**Air-hardening or Self-hardening Steels**—Chromium air hardening steels have found some application in mild high speed cutting service and have in certain cases successfully replaced the tungsten high speed steels. While their red hardness is far from being equal to that of the high tungsten steels, they are nevertheless satisfactory for many purposes.

These steels generally contain between 12 and 17 per cent chromium and from 1.00 to 2.25 per cent carbon. In addition, some of these steels contain up to 3.50 per cent cobalt. The steels are hardened by quenching at about 1800 degrees Fahr. into oil or air, and tempering at a low temperature, 300 to 500 degrees Fahr. They may also be treated to develop high secondary hardness by quenching at 2100 degrees and tempering at 1000 degrees Fahr., but the first is the treatment generally used.

The following Rockwell-C hardnesses are developed when such steels are quenched and tempered as shown.

Tempered Degrees Fahr.	Quenched 1800 Degrees Fahr.	Quenched 2100 Degrees Fahr.
300	66	43
500	65	43
700	63	43
900	62	44
1000	61	45
1100	57	65
	55	62

**High-Speed Steels**—Chromium has been an essential constituent of tungsten high speed steels for many years. In the 12 per cent and 18 per cent tungsten steels, chromium is used almost universally to the extent of 4 per cent, though the proportions range as low as 3.0 per cent and as high as 5.5 per cent. It promotes toughness in high speed steel by entering into solid solution in the matrix, contributes to the hardness by becoming part of the carbides, and makes its presence manifest in the phenomena of secondary hardness. It contributes to the ease of manufacturing by promoting the solution of the tungsten carbides in the solid steel, thus promoting uniformity. When present in amounts over 4 per cent, it makes the high speed steel noticeably stiffer when hot, thus rendering it more difficult to roll and hammer; the same circumstance would, however, contribute to red hardness and therefore increase cutting efficiency in the finished cutting tools. High chromium is also said to increase the danger of grinding cracks in grinding the hardened tools.

**Magnet Steels**—The comparatively inexpensive chromium magnet steels have replaced to a considerable extent the tungsten magnet steels used formerly. A large proportion of these chromium steels have the following chemical composition:

C	Mn	Cr
.95	.45	2.25

The chromium may range as high as 4 per cent, but the other elements remain about as shown.

The steel is hardened by quenching from about 1525 degrees Fahr. into oil. Tempering is usually dispensed with unless it be an artificial aging by heating in boiling water. The steel with this treatment shows a remanence,  $B_r$ , of about 10,000 and a coercive force,  $H_c$ , of about 55 to 63. This compares favorably with tungsten magnet steels which show a remanence of about 9,000 and a coercive force of about 65 to 70. If the steel has had to be annealed for cold shearing or machining, the magnetic values developed after hardening are found to be lower.

**Miscellaneous**—Chromium finds application as a constituent of many other steels, such as rolls for cold rolling, with 2 per cent chromium, semi-hot die steels with 4 per cent chromium, drill steels, chisel steels, and file steels with 0.50 per cent chromium, and numerous special steels requiring high hardness.



# CONSTITUTION OF NICKEL-CARBON ALLOYS

BY N. B. PILLING\* AND T. E. KIHLGREN\*

Carbon has a limited solubility in molten nickel. The equilibrium involved has been determined as to its essential points by Ruff and Borman<sup>4</sup> and more recently confirmed by Kasé,<sup>2</sup> who has added further details regarding the solubility of carbon below the eutectic temperature. The diagram shown in Fig. 1 is that of Kasé, determined with nickel containing about 1 per cent of impurities other than carbon. The impurities were mostly cobalt and iron.

The addition of carbon to nickel progressively lowers the melting point from 1452 to 1318 degrees Cent., at which temperature a eutectic is formed with 2.2 per cent carbon, the carbon appearing as graphite. At the eutectic temperature 0.55 per cent carbon is retained in solid solution, and the amount thus retained decreases with falling temperature to probably less than 0.25 per cent at room temperature. The Bureau of Standards reports<sup>3</sup> a solid solubility of less than 0.1 per cent in a heat-treated nickel containing 2 per cent carbon.

The equilibrium for carbon concentrations greater than 2.2 per cent has not been satisfactorily worked out and data are meager. In 1912 Ruff and Martin<sup>7</sup> measured the solubility of carbon in nickel at high temperatures and obtained the following results:

Temperature Degrees, Cent.	Total Carbon Dissolved Per Cent
1560 .....	2.50
1640 .....	2.69
1640 .....	2.73
1744 .....	2.98
1842 .....	3.38
1950 .....	4.04
1956 .....	3.97
2050 .....	5.30
2100 .....	6.42
2200 .....	6.38
2200 .....	6.30
2300 .....	6.32
2400 .....	6.31
2500 .....	6.30

With these results, the thermal data of Kasé,<sup>2</sup> which extend up to 1700 degrees Cent., are in good agreement. These data have been interpreted to indicate the existence at these high temperatures of a carbide of nickel,  $\text{Ni}_3\text{C}$ , containing 6.38 per cent carbon. If such is the case it must suffer decomposition to graphite and nickel, as it does not occur as a constituent in nickel. Meyer and Scheffer<sup>1</sup> have recently synthesized a carbide of nickel at 270 degrees Cent. which had a decomposition temperature below 700 degrees Cent.

Carbon evidently has little effect on the vapor pressure of nickel, because nickel containing 6.3 per cent carbon boils under 30 millimeters

\*The International Nickel Co., Inc., Bayonne, N. J.

Prepared for the Nonferrous Data Sheet Committee of the Institute of Metals Division of the A. I. M. E. and the Recommended Practice Committee of the A. S. S. T.

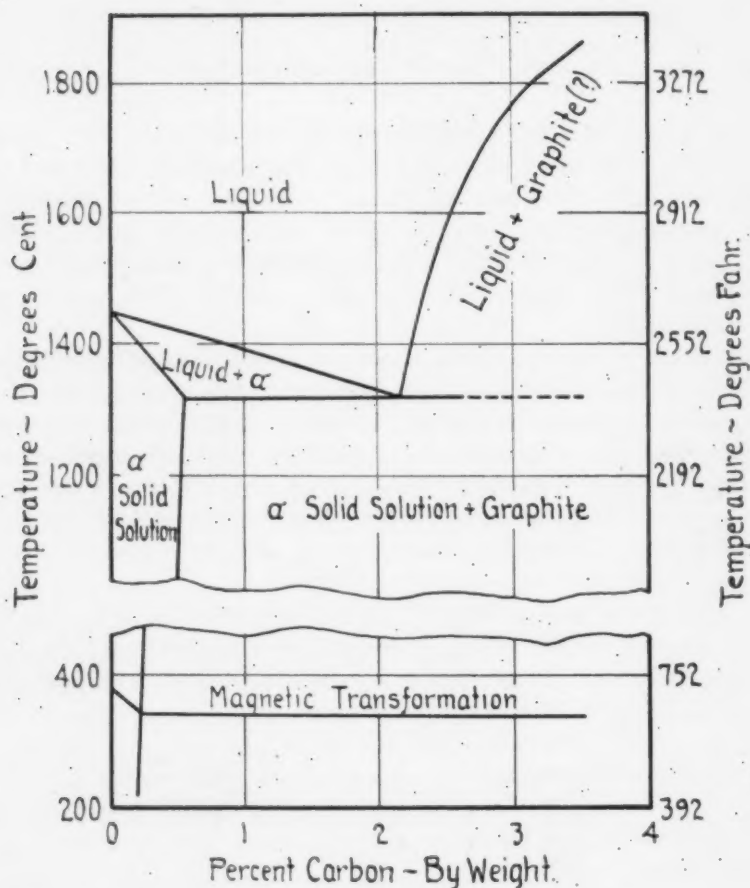


Fig. 1—Nickel-Carbon Equilibrium Diagram (Kasé).

pressure at 2490 degrees Cent., while pure nickel under similar conditions boils at 2400 degrees Cent.<sup>4</sup>

The magnetic transformation of nickel at 357 degrees Cent. is lowered appreciably by carbon in solid solution.<sup>2</sup> As far as is known, no crystallographic change occurs at the magnetic transformation, and both magnetic and non-magnetic conditions have a face-centered cubic structure.

Since nickel exists in but a single phase, carbon brings about no effect in nickel analogous to that which it induces in iron, and consequently the nickel-carbon alloys are not subject to profound structural and physical changes such as occur in steel.

#### REFERENCES

1. Meyer and Scheffer, *Recueil des Travaux Chimiques des Pays Bas*, Vol. 44, 1927, p. 1.
2. †T. Kasé, *Science Reports*, Tohoku Imperial University, I, Vol. 14, 1925.
3. Bureau of Standards Circular No. 100, 1924, p. 40.
4. †O. Ruff and W. Bormann, *Zeitschrift für Anorganische Chemie*, Vol. 88, 1914, p. 386.
5. O. Ruff and E. Gersten, *Berichte*, Vol. 46, 1913, p. 400.
6. O. Ruff, *Berichte*, Vol. 45, 1912, p. 3139.
7. †O. Ruff and W. Martin, *Metallurgie*, Vol. 9, 1912, p. 143.
8. †K. Friedrich and F. Leroux, *Metallurgie*, Vol. 7, 1910, p. 101.

†These are the most important.

# THE ENGINEERING INDEX

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In the preparation of the Engineering Index by the staff of the American Society of Mechanical Engineers some 1700 domestic and foreign technical publications received by the Engineering Societies Library (New York) are regularly searched for articles giving the results of the world's most recent engineering and scientific research, thought, and experience. From this wealth of material the A. S. S. T. is supplied with this selective index to those articles which deal particularly with steel treating and related subjects.

## AERONAUTICAL RESEARCH

Aeronautical Research at the National Bureau of Standards, L. J. Briggs. *Aero Digest*, vol. 14, no. 3, Mar. 1929, pp. 47-49, 5 figs.

Investigations being carried out by Bureau in co-operation with (and financial assistance of) Aeronautics Branch of Department of Commerce, National Advisory Committee for Aeronautics, Bureau of Aeronautics of Navy, and Army Air Corps; directive radio beacon system; relative visibility of neon lamp and incandescent lamp beacon in fog; sound-proofing airplane cabins; wind tunnel research; aircraft instruments; corrosion of duralumin.

## AIRPLANE FRAMES

STATIC TESTING. Static Testing of Airplanes, J. S. Newell. *Airway Age*, vol. 10, no. 5, May 1929, pp. 619-622, 4 figs.

Description of 11 tests which are necessary for proper determination of structural strength of airplane frames; for wing ribs two conditions of loading should be considered; testing tail surfaces; overload on control system; testing media; ailerons test; report should include method.

STRESSES. Development of Metal Construction, H. J. Pollard. *Flight (Lond.)*, vol. 21, no. 17, Apr. 25, 1929, (*Aircraft Engineer sec.*) pp. 25-28, 6 figs.

Calculations of moments of inertia of beams made up of corrugated strip of small thickness; moment of inertia of flange about central axis of spar; moment of inertia of web about same axis obtained in similar manner, thus giving total moment of inertia of section; fiber stresses in spar when given loads come on to it can be determined; stress at which elastic instability is likely to set in must be estimated for these flanges and webs.

STRESSES. Experimental Determination of the Stresses in the Framework of Airplanes (Determination experimentale des tensions dans les charpentes d'avions), W. Margoulis. *Aéronautique (Paris)*, vol. 11, no. 118, Mar. 1929, pp. 81-83, 3 figs.

Description of apparatus developed for testing airplane frames, similar to photoelasticity method developed by Mesnager, and based upon measure of strain optically by photo-

elasticity method; tests on Gourdore and Le-seurre pursuit plane described; principles of apparatus employed for making photoelastic measurements are explained.

## AIRPLANE MANUFACTURE

Cardinal Factors in Aircraft Production, W. Nelson. *Airway Age*, vol. 10, no. 5, May 1929, pp. 638-641, 4 figs.

Description of principles employed in mass production of airplanes which closely parallel those used in automotive industry; to obtain low cost in producing certain units, plant, machines, raw material, and labor must be coordinated into pre-determined scheme of production; importance of straight-line production; use of jigs for assembly; function of machine tools.

Construction of Modern Airplanes (La construction des avions modernes), C. Martinot-Lagarde. *Pratique des Industries Mécaniques (Paris)*, vol. 11, no. 11, Feb. 1929, pp. 441-453, 20 figs.

Use of metal construction in airplanes is discussed; comparison of metal and wood construction; steel and duralumin used; hardening temperature; mechanical operations including setting, cutting out plates, swaging, plate bending, drawing, and riveting; preservation against oxidation; general design and different system of metal construction of airplanes in various countries described.

BRAZING. Silver Brazing and Soldering in Airplane Construction. *Aero Digest*, vol. 14, no. 3, Mar. 1929, p. 96.

Use of silver solders by various manufacturers in brazing parts of planes or engines that are subject to vibration or likely to encounter relatively high operating temperatures, is described; United States Army specifications.

METAL CONSTRUCTION. Metal Aircraft Construction, R. Merchant. *Aviation Eng.*, vol. 2, no. 5, May 1929, pp. 7-8 and 13, 3 figs.

Description of patented fabricated metal wing structure used by Nicholas-Beazley Airplane Co. for Barling NB-3 monoplane; every part is made by machinery designed and perfected for that purpose; entire struc-

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ture made up of one principal U-shaped box spar of stamped duralumin parts riveted together; seven different thicknesses of duralumin wing structure; fuselage construction.

UNITED STATES. "The Infant Industry," F. D. Walton. *Aero Digest*, vol. 14, no. 4 Apr. 1929, pp. 58-66, 58 figs.

Facts revealed by study of American aircraft production today; no small degree of confusion exists; small conservative group is trimming ship for bad blow through which it seems inevitable aviation must pass in future; reasons for slight loss in production in New York; airplane manufacture is discussed by sections of country.

WELDING. Welding Practice in the Aircraft Industry, J. F. Hardecker and E. A. Joyce. *Welding Engr.*, vol. 14, no. 4, Apr. 1929, pp. 43-45, 10 figs.

Actual shop processes of oxyacetylene welding used in manufacture of airplanes are described; selection of equipment; materials used in welding and welding rod recommended; different types of welded joints are shown; aircraft welding requirements; procedure control; duties of supervisor.

#### AIRPLANE MATERIALS

Airplane metals (Die Metalle des Flugzeugbaus), J. Hausen. *Zeit. für Flugtechnik und Motorluftschiffahrt (Munich)*, vol. 20, no. 4, Feb. 28, 1929, pp. 92-94.

Abstracts of papers and discussions on characteristics of metals and alloys, presented at special session of German Metallurgical Society.

#### AIRPLANES

DYNAMIC STRESSES. The strength of Aircraft Members, P. Brenner. *Metallurgist (Supp. to Engineer, Lond.)*, Apr. 26, 1929, pp. 52-53.

Discussion of various factors which may give rise to alternating or fatigue stresses in airplane; main causes are aerodynamic and engine effects; parts most affected by engine vibration are engine bearers, radiator and fuel piping and controls; fundamentals of fatigue failure; examples; recommendations as to procedure in order to ensure that all aircraft members shall have sufficient strength against alternating stresses. Abstract-translated from V.D.I. Zeit. Dec. 22, 1928.

#### ALLOY STEEL

Recent Developments in Alloy Steels, R. Hadfield, W. B. Pickering and S. A. Main. *Iron and Steel of Canada (Gardenvale, Que.)*, vol. 12, no. 4, Apr. 1929, pp. 105-108, 10 figs.

Noteworthy technical advances in alloy steels, especially in development of heat and corrosion-resisting steels, and successful industrial applications are outlined; high-percentage nickel-chromium steels; microstructure of Era heat-resisting steel and strength at high temperatures; testing apparatus for finding limiting creep stress of Era steel.

ANALYSIS. Determination of Chromium, Tungsten, Molybdenum, Vanadium, Nickel, Manganese, and Cobalt in High-Alloy Steels (Bestimmung von Chrom, Wolfram, Molybdaen, Vanadium, Nickel, Mangan und Kobalt in hochlegierten Stählen), H. Mende. *Chemiker-Zeitung (Koethen)*, vol. 53, no. 18, Mar. 2, 1929, pp. 178-179.

Detailed directions are given for analysis

of alloy steels containing high percentages of metals other than iron.

PROPERTIES. Welding Facts and Figures, D. Richardson and E. W. Birch. *Welding J. (Lond.)*, vol. 26, no. 306, Mar. 1929, pp. 70-73.

Properties of acid and heat-resisting alloys and high-speed steel are discussed; air-hardening properties of high-speed steel; procedure for forging, annealing, hardening, and secondary hardening. (Continuation of serial.)

TEMPERING CHANGES. Tempering Changes in Steels, R. Hay and R. Higgins. *Roy. Tech. College—Jl. (Glasgow)*, vol. 2, part 1, Jan. 1929, pp. 73-80, 5 figs.

Study of these changes has been extended to alloy steels.

#### ALLOYS

ANNEALING. The effect of Annealing Upon the Solidus Temperature of Alloys, J. H. Andrew and H. M'Neil. *Roy. Tech. College—Jl. (Glasgow)*, vol. 2, part 1, Jan. 1929, pp. 64-72, 5 figs.

Experiments carried out with copper-tin alloys giving normally range of 150 degrees Cent.; question of grain refinement is dealt with and new hypothesis put forward to explain fact that refinement is never produced by cooling through but only by heating through transformation.

ELECTRIC RESISTANCE. A New Method of Measuring the Electrical Resistances of Alloys, A. L. Norbury. *Lond., Edinburgh and Dublin Philosophical Mag. and Jl. of Science (Lond.)*, vol. 7, no. 44, Apr. 1929, pp. 662-669, 4 figs.

Description of method for measuring specific electrical resistance of alloys of any size or shape, provided plane surface about 5 by 5 mm. or larger is available; investigation was commenced with idea of measuring diameters of Brinell impressions while load was on by measuring electrical resistance of circuit comprising Brinell ball and specimen under test.

#### ALUMINUM

COLD-HARDENING. Conjugated Influence of Velocity of Deformation and Temperature in Cold-Hardening Process of Aluminum (Influences conjuguées de la vitesse de déformation et de la température sur la production de l'écrouissage), A. Roux and J. Cournot. *Académie des Sciences—Comptes Rendus (Paris)*, vol. 188, no. 9, Feb. 25, 1929, pp. 631-633.

Aluminum is cold-hardened by instantaneous rupture at 380 to 450 degrees Cent. and metal exists in hardened state above normal annealing temperature; for each temperature there is velocity of deformation above which simultaneous recrystallization does not immediately annul hardening effect of deformation.

#### ALUMINUM ALLOY CASTINGS

X-RAY ANALYSIS. Radiography of Light Metal Alloys, S. H. Anderson and L. Marick. *Univ. of Wash. Eng. Experiment Station—Bul.*, no. 49, Dec. 1, 1928, 18 pp., 7 figs.

Work was undertaken for purpose of determining under what conditions of voltage, tube current, and exposure best contrast is obtained for particular castings and shapes radiographed; effect of various factors involved in technique of radiography are



brought out; X-Ray tube used was Coolidge, tungsten target tube of radiator type, maximum allowable current and voltage being 30 milliamperes and 85 kilovolts respectively.

#### ALUMINUM ALLOYS

**TERNARY.** Ternary Equilibria, Hamasumi and Matoba. *Metallurgist (Supp. to Engineer, Lond.)*, Apr. 26, 1929, pp. 53-54, 2 figs.

Solution is submitted of ternary systems based on same principle which underlies that of binary equilibria, i. e., that equilibria diagrams may be regarded as assemblage of solubility curves whose intersections are controlled by phase rule; authors apply their method of analysis to aluminum-copper-zinc system. Abstract from Tohoku Imperial University—Technological Reports, vol. 8, no. 1, 1928.

#### ALUMINUM CASTINGS

Aluminum Castings, E. T. Painton, *Elec. Rev. (Lond.)*, vol. 104, no. 2683, Apr. 26, 1929, pp. 731-734, 6 figs.

Discussion of that part of electric industry which is concerned with machine construction, for which aluminum in form of castings is possible, constructional material; various parts used in railway and electric industry are illustrated; physical properties of chill-cast aluminum alloys; comparative properties of aluminum brass and iron.

Advantages of Aluminum as a Foundry Material, G. Mortimer. *Can. Foundrymen (Toronto)*, vol. 20, no. 4, Apr. 1929, pp. 21-24.

Purchase of materials is discussed; scrap should be purchased carefully; saving is small; adding zinc and copper; melting aluminum and copper in separate crucibles; main essentials of melting practice; absorption of gases; reverberatory furnaces; pour temperatures.

**SOLDERING.** Principles in Soldering Aluminum Castings. *Airway Age*, vol. 10, no. 5, May 1929, p. 629.

Precautions to be observed in soldering aluminum castings are discussed.

#### ALUMINUM CORROSION

Corrosion of Aluminum (Sur la corrosion de l'aluminium), J. Calvet. *Académie des Sciences—Comptes Rendus (Paris)*, vol. 188, no. 17, Apr. 22, 1929, pp. 1111-1114.

Results of acid-corrosion tests of extra pure, Hoopes and ordinary commercial aluminum, showing great superiority of pure aluminum in resisting corrosion.

#### ALUMINUM WELDING

Abstract of Paper on Welded Aluminum in the Chemical and Process Industries, W. M. Dunlap. *Am. Welding Soc.—Jl.*, vol. 8, no. 4, Apr. 1929, pp. 13-20, 7 figs.

In comparison with some other corrosion-resistant metals and alloys, aluminum has advantage even from standpoint of initial cost; resistance to corrosion and chemical action; no other metal should be in contact with aluminum and with corrosive agent; strength of welds; summary of tensile tests; practically no difference between tensile strength of welded tubes and original tubes. Abstract.

#### AUTOMOBILE MANUFACTURE

**AXLES.** Axle Gears and Housings, R. C. Wilson. *Soc. Automotive Engrs. Jl.*, vol. 24, no. 5, May 1929, pp. 530-531.

Materials, methods and tolerances used and recommended by Gleason Works in manufacture of axle gears and housings are outlined; machining gears and pinions; gear lapping; accuracy required in mountings.

**FORGING.** Automobile Hardware will be Forged, F. W. Manker. *Heat Treating and Forging*, vol. 15, no. 4, Apr. 1929, pp. 484-485, 3 figs.

Details of rolling mills and forging equipment of Cleveland Hardware Co., employed in manufacture of automobile hardware; heat treating equipment; efficient combustion system.

#### BRIDGE WIRE MANUFACTURE

Manufacture of Bridge Wire, E. E. Thum. *Iron Age*, vol. 123, no. 19, May 9, 1929, pp. 1281-1285, 4 figs.

Description of processes used by John A. Roebling's Sons Co. to manufacture bridge wire for mammoth suspension bridge now being built across Hudson River at New York; wire is high-carbon acid open-hearth steel made of boiler plate scrap and manganese-bearing pig iron; manufacture of acid steel; blooming and billet mills; manufacture of rods; preparation for wire drawing; wire drawing operation; galvanizing and inspection.

#### BRIDGES, SUSPENSION CABLES

**DETROIT RIVER.** Find Breaks in Detroit Cables. *Eng. News-Rec.*, vol. 102, no. 20, May 16, 1929, p. 808.

Dismantling work on Ambassador suspension bridge at Detroit has brought to light numerous broken cable wires located beneath cable wrapping at points of suspender connections.

#### CASE HARDENING

**DEPTH MEASUREMENT.** Case Depth Measurement. *Am. Soc. Steel Treating—TRANS.*, vol. 15, no. 4, Apr. 1929, pp. 729-732.

Recommended practice committee release; methods for measuring case depth include scale measurement of ordinary fracture, of blued fracture, of ordinary fracture which has been dipped in acid, and of polished cross sections; microscopic measurement of highly polished cross sections; file or drill through from core at right angles to case until hardness of case resists further cutting or abrasion; magnetic analysis; etc.

#### CAST IRON

**ALLOY.** Effects of Nickel and Chromium on Cast Iron, D. Hanson. *Metallurgist (Supp. to Engineer, Lond.)*, Apr. 26, 1929, pp. 56-60, 9 figs.

Effects of nickel and silicon when present together; effect of chromium; comparison of effects of silicon, nickel, and chromium; practical applications.

**ANNEALING.** The Annealing of Cast Iron, H. H. Beeny. *Foundry Trade Jl. (Lond.)*, vol. 40, nos. 658 and 659, Mar. 28 and Apr. 4, 1929, pp. 229-231, 8 figs.

Account of author's experiments performed in 1925; irons were cast in green sand, and were made into quite small pieces for Brinell

and other tests; temperature of 700 degrees Cent. was found slow in eliminating all combined carbon; tests show temperature, comparative stability of combined carbon at temperature over 800 degrees Cent. and relation of its amount to silicon content; it is shown that indiscriminate annealing of gray-iron castings is thoroughly bad.

#### CHROMIUM STEEL

High Chromium Steels, O. K. Parmitter. *Am. Soc. Steel Treating*—TRANS., vol. 15, no. 5, May 1929, pp. 796-816.

Composition, heat treatment, and properties of high-chromium steels, including various low-carbon, stainless steel types and several high carbon-chromium die types discussed in general; special consideration is given to more recent developments of super-stainless steel; effect of various chemicals and other corrosive substances upon stainless steel is given in detail.

#### CHROMIUM-ALUMINUM STEEL

An investigation of the Physical Properties of Certain Chromium-Aluminum Steels, F. B. Lounsberry and W. R. Breeler. *Am. Soc. Steel Treating*—TRANS., vol. 15, no. 5, May 1929, pp. 733-762 and (discussion) 762-766, 13 figs.

Physical properties investigated for steel containing as main alloying constituents, aluminum and chromium with silicon and carbon; aluminum effectively stabilized alpha-iron phase and when in combination with chromium, greatly increased resistance of metal to oxidation at elevated temperatures; addition of nickel produced unusual low-temperature phenomenon.

#### CHROMIUM-NICKEL STEEL

Investigations on Creep Strength, H. J. Tapsell and J. Remfry. *Iron and Coal Trades Rev. (Lond.)*, vol. 118, no. 3191, Apr. 26, 1929, p. 605, 3 figs.

Tabular data on results of short-time tensile tests at air temperature, 500 degrees Cent., 600 degrees Cent., 700 degrees Cent. and 800 degrees Cent.; creep tests at 600, 700, and 800 degrees Cent. Abstract of Department of Sci. and Ind. Research—Special report, no. 15.

TESTING. The Creep Strength of a High Nickel High Chromium Steel, Between 600 degrees Cent. and 800 degrees Cent. H. J. Tapsell and J. Remfry. *Dept. Sci. and Indus. Research—Report (Lond.)*, no. 15, 1929, 7 pp. 7 figs.

Results of tensile and creep tests of forged bar 4 ft. long by 2 in. sq. finished at temperature of 950 degrees Cent. consisting of 26.5 per cent nickel, 14.0 per cent chromium, and 3.59 per cent tungsten.

#### COPPER ALLOY CASTINGS

Some Notes on Non-Ferrous Casting Alloys, W. Lambert. *Mech. World (Lond.)*, vol. 85, no. 2203, Mar. 22, 1929, pp. 275-276.

Copper-aluminum bronzes and copper-tin group of alloys, true bronzes, are discussed, latter group including as it does various types of red brass, gunmetal, and phosphor-bronze alloys; admiralty gunmetal best for gunmetal castings; properties of phosphor and aluminum bronzes; defective castings. Paper read before Inst. of British Foundrymen.

#### COPPER ANNEALING

The Reduction of Cuprous Oxide by Carbon Monoxide, C. G. Maier. *U. S. Bur. of Mines—Reports of Investigations*, no. 2926, Apr. 1929, 7 pp.

Paper deals with calculation of gas concentrations in reduction of cuprous oxide, and refers especially to conditions of temperature and concentration; present wire drawing practice often involves use of steam for protection during annealing, and it is not impossible that substitution of carbon dioxide containing small amounts of carbon monoxide might result in considerable saving.

#### CRYSTALS—STRUCTURE

Crystal Structure Types, G. W. Gruner. *Am. Mineralogist*, vol. 14, no. 5, May 1929, pp. 173-187, 17 figs.

Much of information collected in recent years on structure of crystals is not in domain of mineralogy, but large part of it is becoming of importance to this science; atomic arrangements and type structures; morphotropism and polymorphism; structures containing radicals. Bibliography.

#### DROP FORGINGS

HEAT TREATMENT. Heat treatment of Drop Forgings, C. L. Foreman. *Fuels and Furnaces*, vol. 7, no. 4, Apr. 1929, pp. 497-500, 3 figs.

Heat treatments of drop forgings in continuous furnace with automatic temperature control insures uniformity of product, as conditions of heating, time at heat, and weight of steel treated per hour, may be duplicated day after day; normalizing; annealing and hardening treatments.

#### ELECTRIC FURNACES

ANNEALING. Technological Study of the Bright Annealing of Steel in the Electric Furnace (Technologische Studien ueber das Blankgluehen von Stahl im Elektrischen Ofen), A. Pomp and L. Walther. *Mitteilungen aus dem Kaiser-Wilhelm-Institut fuer Eisenforschung (Duesseldorf)*, vol. 11, no. 2, 1929, pp. 15-30, 60 figs. partly on supp. sheet.

Construction and operation of 25-kw. annealing furnace is described; by annealing in electric furnace, properties of deep-drawing strip steel were improved in comparison with those of material treated in coal-fired pot annealing furnace, owing to better control over heat distribution, temperature, and duration of annealing; tests showed that hydrogen-filled electric furnace was suitable for recrystallization anneals.

ANNEALING. Electricity for Normalizing and Annealing, J. R. Miller. *Heat Treating and Forging*, vol. 15, no. 4, Apr. 1929, pp. 481-483, 8 figs.

Efficient and economical preparation of alloy steel bars and tubes is accomplished in 14 electric furnaces of special design at Canton, Ohio, plant of Timken Steel and Tube Co.; details of 850-kw. pit annealing furnace; two special 350-kw. car type furnaces.

FERROALLOY MANUFACTURE. The Manufacture of Ferro-Silicon in the Electric Furnace, J. Silberstein. *Can. Chem. and Met. (Toronto)*, vol. 13, no. 4, Apr. 1929, pp. 91-94.

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Account of furnace design, construction and operation; power consumption in 2000 to 4000-kw. furnaces ranges from 2.1 to 2.6 kw-hr. per lb. 50 per cent ferrosilicon when coke is used as reducing agent; electrode consumption per 100 lb. produced ferrosilicon varies from 3 to 6 lb. with amorphous carbon, and 2 to 4 lb. with Soderberg self-baking electrodes.

**HEAT TREATING.** Progress Made in the Use of Electric Furnaces for Heat Treating, A. N. Otis. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 5, May 1929, pp. 767-794 and (discussion) 794-795, 13 figs.

Present status of electric furnaces, extent of their use for heat treating, and some new types that have been recently developed; advantages for heat treating; some of individual furnaces and their operation briefly described; electric furnaces for forging; continuous furnaces for new process of copper brazing, and for annealing in atmospheres of protecting gas.

**HEAT TREATING.** Electric Heat Treating Benefits Evaluated, W. S. Scott. *Elec. World*, vol. 93, no. 19, May 11, 1929, pp. 929-933, 7 figs.

Economical cost analysis of electric heat treatment of small tools; conditions on which analysis is based; improvement in quality; dependable and increased production; decrease in labor cost; better working conditions; ease of operation; net value equals per cent return on investment for specific case; five advantages individually offset 30 to 128 per cent of fuel cost.

**HEAT TREATING.** Heat-Treating Aluminum Alloys, F. L. Faurote. *Iron Age*, vol. 123, no. 20, May 16, 1929, pp. 1347-1348, 2 figs.

Description of electric furnaces in plant of Wright Aeronautical Corp., Paterson, N. J., which are employed in treating aluminum alloy parts for airplane engines; five furnaces of two sizes; cover-lifting mechanism; four large furnaces have three zones.

**HEAT TREATING.** Heat on a Quality Basis, E. J. Cipperly. *Elec. World*, vol. 93, no. 17, Apr. 27, 1929, pp. 827-828, 1 fig.

Demand for uniformly high-grade product leads Los Angeles manufacturer of oil-well tools to install five electric heat-treating furnaces; three 45-kw. box-type units are used for treating small parts; two car-type units operated on 24-hr. heat-treating cycle practically without attendance.

**HIGH-FREQUENCY.** Progress in Design of High-Frequency Furnace Installations (Fortschritte im Bau von Hochfrequenz-Ofenanlagen), M. Tama. *Stahl und Eisen (Duesseldorf)*, vol. 59, no. 15, Apr. 11, 1929, pp. 499-502, 7 figs.

Basic principles and developments in past few years; description of coreless induction furnace; electrical conditions in melting of magnetic and non-magnetic metals; general arrangement of high-frequency furnace; production of crucibles; electric current consumption.

**METALLURGICAL.** Electric Furnaces. *Metallurgist (Supp. to Engineer, Lond.)*, Apr. 26, 1929, pp. 50-51.

It has come to be recognized that electric furnace in nonferrous metal melting and annealing has established important place for itself; way in which application of electric furnaces affects conditions of work in nonfer-

rous industries; there need be no dirt or dust and very little waste heat.

**RESISTANCE.** Industrial Electric Heating, N. R. Stansel and C. Dantszen. *Gen. Elec. Rev.*, vol. 32, nos. 3 and 5, Mar. and May 1929, pp. 162-169 and 278-283, 12 figs.

Mar.: Resistor furnaces; atmospheres in heating chambers, and classification of gas atmospheres on basis of material to be heated and process in view; Dallon's law and its application; dissociation of oxides. May: Furnace atmospheres and carburizing reactions in iron and steel are discussed.

**STEEL MAKING.** Coreless Induction Furnaces. *Metropolitan-Vickers Gaz. (Manchester)*, vol. 11, no. 192, Mar. 1929, pp. 228-231, 5 figs.

Electric steel-melting furnaces of Metropolitan Vickers Co. of 5 cwt. and 350 lb.; electric motor-generator equipment for 5-cwt. type of 160 kw. capacity, 1000 volt 500 cycles; performance data.

## ELECTROPLATING—ALUMINUM

The Testing of Electro-Deposits on Aluminum, G. B. Brook and G. H. Stott. *Engineering (Lond.)*, vol. 127, no. 3302, Apr. 26, 1929, pp. 531-534, 6 figs.

General features of, and reasons put forward to account for, unsuccessful deposits on aluminum; character and thickness of deposits; soundness of deposits; tests for soundness of adhesion; corrosion tests; while any electrodeposit should be of such reasonable thickness as to afford wear, its resistivity to corrosion and its strength of adhesion are dependent upon physical character of coating equally with thickness of deposit. Paper read before Inst. of Metals.

## FURNACES

**CONTINUOUS HEATING.** Development of the Continuous Pack-Heating Furnace, J. A. Hunter. *Engrs. Soc. of West. Pa.—Proc.*, vol. 45, no. 1, Feb. 1929, pp. 1-9, 7 figs.

Brief description of old and new methods of rolling thin sheets; principal furnace requirements; details of design of furnace 30 ft. long by 42 in. wide consisting of revolving alloy disks carried on 32 shafts.

**HEAT TREATING, GAS FIRED.** Design and Operation of Gas-Heated Furnaces, C. M. Walter. *Gas J. (Lond.)*, vol. 186, no. 3439, Apr. 17, 1929, p. 151.

Author points out that in some of more modern designs of recuperators, heat-resisting alloys such as nickel chrome have been used for flues with considerable advantage; first essential for efficient working of gas furnace is that combustion shall be as complete as possible; losses by radiation and convection can be cut down by paying every attention to lagging of outside walls and providing well fitting doors.

**METALLURGICAL—REFRACTORY MATERIALS.** Refractories and Heat Insulation Connection with Furnace Design, G. A. Barker. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 5, May 1929, pp. 851-858. See also *Fuels and Furnaces*, vol. 7, no. 4, Apr. 1929, pp. 611-612.

Change in furnace design from great masses of brickwork to modern thin-walled heating machines is pictured; thin wall of new insulating materials will transmit less heat than very much thicker wall of solid refractory material; refractory cements and their function in cutting down heat losses and protect-



ing refractory linings are quite thoroughly discussed.

#### IMPACT TESTING— NOTCHED BAR

Notched-Bar Test, Its Development and Appraisal (Die Kerbschlagprobe Entwicklung und Kritik), F. Fettweis, *Archiv für das Eisenhüttenwesen (Düsseldorf)*, vol. 2, no. 10, Apr. 1929, pp. 625-674, 31 figs.

Historical review; machines, apparatus, and methods employed; determination of deformation and fracture phenomena; tests to explain reason for different kinds of fracture; notched-bar test and law of similitude; relation of impact to temperature; relation of notch hardness to structure, heat treatment, and chemical composition; shortcomings of notched-bar test; suggestions for improvement. Includes 12-page bibliography and subject index.

#### IRON AND STEEL

Sound Ingots Are Under Study. *Iron Age*, vol. 123, no. 20, May 16, 1929, pp. 1367-1368.

Review of meeting of British Iron and Steel Institute, with brief abstracts of papers; New Method for Production of Sound Steel, C. A. Parsons; report of committee on heterogeneity of steel ingots; report of committee on trends in blast-furnace practice; method of drying atmospheric air by means of "silica gel" and installation of such plant at Scottish works, E. H. Lewis.

**OXIDATION.** The Oxidation of Iron and Steel at High Temperatures, L. B. Pfeil. *Iron and Steel Inst.—Advance paper (Lond.)*, no. 11, May 1929, 47 pp., 20 figs. See also abstract in *Iron and Coal Trades Rev. (Lond.)*, vol. 118, nos. 3193 and 3194, May 10, and 17, 1929, pp. 717-718 and 750-752.

Paper deals with mechanism of oxidation, and not with rate of oxidation in quantitative sense; appearance, composition, and general properties of scale; mechanism of oxidation; cause of layer formation; envelopment of foreign particles; crystalline scale deposits; outer part of scale is last to form, and middle portion first to form. Bibliography.

**PROPERTIES.** The Change of Specific Gravity of Cold-Worked Iron and Steel by Tempering, K. Tamaru. *Inst. Phys. and Chem. Research—Bul. (Tokyo)*, vol. 8, no. 3, Mar. 1929, pp. 25-26.

Temperature of strain release for iron and steel was determined from change of specific gravity, and was found to be 400 degrees Cent. for all specimens treated; when strained specimen is tempered at low temperature minute increase in grain size takes place and specific gravity increases to yield maximum; if temperature is raised still higher, interior of grain separates into series of thin layers, and increases total surface of grain, thus yielding minimum density.

#### IRON AND STEEL— SULPHUR DETERMINATION

Influence of Copper on Sulphur Determination in Iron and Steel (Ueber den angeblichen Einfluss des Kupfers auf die Schwefelbestimmung im Eisen und Stahl), H. Pinski. *Glaser's (Düsseldorf)*, vol. 16, no. 20, May 17, 1929, pp. 453-458, 3 figs.

Reference is made to supposedly low sulphur content in specimens containing cop-

per; copper sulphide in polished specimen; phenomena occurring with dissolution in hydrochloric acid; copper determination.

#### IRON FOUNDRY PRACTICE

Gray Iron Electrically Melted. *Iron Age*, vol. 123, no. 20, May 16, 1929, pp. 1351-1353, 4 figs.

Description of small iron foundry of Hobart Brothers Co., Troy, Ohio, having daily output of two tons, which uses method commonly associated with mass production; iron melted in rotating electric furnaces; furnace lined with special material; sand handled as in larger foundries; castings sprayed with paint; sand-reclamation equipment designed by company.

#### IRON-NICKEL ALLOYS

Constitution of Iron-Nickel Alloys, P. D. Merica. *Am. Soc. Steel Treating—TRANS.*, vol. 15, no. 5, May 1929, pp. 881-884, 1 fig.

Recommended Practice Committee release; irreversible alloys are subjected to heat treatment in same sense and same manner as steel; segregate of alpha phase resulting from its separation from gamma resembles structure and mode of formation of martensite.

#### INDUSTRIAL RESEARCH

Research an Essential Factor in Industrial Progress, B. G. Klugh. *Min. Congress J.*, vol. 15, no. 5, May 1929, pp. 346-348.

perpetual inventory as to production costs, perpetual inventory as to production costs, processes and equipment, available markets, quality of product, competitive products and expansion program; research organization equips executive to obtain knowledge of his business necessary to progress.

#### IRON CORROSION

The Mechanism of Oxidation—Rusting of Iron (Ueber den Mechanismus der Oxydations-Vorgaenge, etc.), H. Wieland and W. Franke. *Annalen der Chemie (Berlin)*, vol. 469, no. 3, Apr. 19, 1929, pp. 257-308.

Report from Chemical laboratory of Bavarian Academy of Sciences on chemistry of rusting; formation of hydroperoxide when iron is rusting, auto-oxidation of iron in alloy and in powdered forms; effect of passivating agents on auto-oxidation; replacing oxygen with other elements readily combined with hydrogen.

#### IRON METALLOGRAPHY

Natural History of the Iron Family (Die Naturgeschichte der Eisenfamilie), V. M. Goldschmidt. *Stahl u. Eisen (Düsseldorf)*, vol. 49, no. 18, May 2, 1929, pp. 601-612, 12 figs.

Properties of elements of the iron family are outlined; peculiarities of atomic structure and its influence on technical evaluation of the elements; quantity and mode of distribution of ferrous metals, and laws regulating these; geochemical behavior; history of origin of primary deposits and formation of sedimentary deposits.

#### IRON AND STEEL

Carbon Content. Determination of Total Carbon in Ferrous Products (La Determinación del Carbono Total en los Productos Siderurgicos), J. Gonzales-Rothvos y Gil. *Quimica e Industria (Barcelona)*, vol. 6, no.

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61, Feb. 1929, pp. 24-30, 6 figs.

Discussion of Fleming gravimetric, Stroehlein volumetric, and Cain-Maxwell electro-metric methods; in each, carbon is burned in electric furnace in current of oxygen, and they differ in manner of measuring carbon dioxide formed; tabular comparative analyses by each method; superiority of Fleming method is indicated; Stroehlein method advantageous for quick results.

**METALLOGRAPHY.** Constitutional Diagrams for Cast Irons and Quenched Steels, A. L. Norbury. *Iron and Steel Inst.—Advance paper (Lond.)*, no. 9, May 1929, 26 pp., 34 figs. See also Abstract in *Iron and Coal Trades Rev. (Lond.)*, vol. 118, no. 3192, May 3, 1929, pp. 667-668 and 673-668 and 673-674, 11 figs.

Diagrams are constructed for iron-carbon alloys containing 0, 2, and 4 per cent of silicon, respectively; normal and supercooled systems are shown for austenite-graphite and austenite-cementite eutectics; in most cases, diagrams discussed have been estimated from existing data.

## MATERIALS

**GRAIN SIZE.** Statistical Description of the Size Properties of Non-Uniform Particulate Substances, T. Hatch and S. P. Choate. *Harvard Eng. School—Publication*, no. 35, 1928-1929, pp. 369-387, 11 figs.

Paper attempts to define particle size by means of well-known statistical methods and to indicate mathematical laws of relationship between particle size, thus defined, and certain size properties of non-uniform particular substances, namely, number of particles per gram of material and specific surface of substance. Bibliography.

## MATERIALS TESTING

**DUCTILITY.** Proposed New Criteria of Ductility from New Law Connecting Percentage Elongation with Size of Test-Piece, D. A. Oliver. *Instn. Mech. Engrs.—Proc. (Lond.)*, no. 4, 1928, pp. 827-850 and (discussion) 850-864, 9 figs.

Description of testing methods and results which indicate that percentage elongation obtained in test is function of original gage length and cross section area; critical examination and analysis of previous investigation.

## METALLOGRAPHY

Metallography Simplified for Practical Use in Shop, E. Preuss, G. Berndt, M. V. Schwarz. *Iron Trade Rev.*, vol. 84, nos. 16 and 18, Apr. 18 and May 2, 1929, pp. 1054-1056 and 1186-1189, 22 figs.

Apr. 18: Structure of crystals in steel is discussed with effects of annealing, overheating, and burning. May 2: Rolling and cold working of steel; with such cold working orientation of crystals also takes place, so that axes arrange themselves in certain location to direction of elongation; tensile strength of iron increases somewhat with grain elongation due to cold working. (Continuation of serial.)

**RESEARCH.** Metals and Atoms. *Metallogrist (Supp. to Engineer, Lond.)*, Apr. 26, 1929, pp. 51-52.

It is claimed that scientific metallurgist is not slow to accord his respect to fundamental discoveries of physicists; but some physicists

appear to know little of what scientific metallurgy has accomplished during past 50 years; study of alloy systems; X-Ray crystal analysis has in main done little more than confirm and amplify inferences arrived at by scientific metallurgists many years ago.

**RESEARCH.** Metallurgy at the National Physical Laboratory. *Chem. Age, (Lond.)*, vol. 20, no. 514, May 4, 1929, pp. 33-34.

Résumé of more important points of report of National Physical Laboratory for year 1928, covering single crystals, light alloys and minor metals; alloys for use at high temperatures, commercial possibilities, and miscellaneous research.

## METALS

**AGE HARDENING.** Hardening. *Metallogrist (Supplement to Engineer)*, Apr. 26, 1929, pp. 49-50.

Notes on dispersion theory of hardening; slip-interference theory of Jeffries and Archer; distortion theory of Rosenhain; news comes from America that attempt, which promises to be successful, has been made to explain on basis of age hardening peculiar phenomena of rehardening which occur when cold-worked iron or steel is heated to moderate temperatures.

**COLD WORKING.** Texture of Cold-Worked Metals (Texturen kaltverformter Metalle) W. E. Schmid. *Zeit. für Technische Physik. (Berlin)*, vol. 10, no. 4, pp. 141-143, 5 figs.

Close relationship of texture of copper, aluminum, and iron after deformation processes, such as stretching, compressing, and rolling are pointed out.

**CORROSION.** Corrosion of Metals (Die Korrosion der Metalle), C. Forehlich. *Gesundheits-Ingenieur (Munich)*, vol. 52, no. 3, Jan. 19, 1929, pp. 33-36, 3 figs.

General review of theories of corrosion and methods of protection against corrosion, based on series of lectures by M. Werner before Technical Scientific Society in Cologne, during winter semester of 1926-27.

**CORROSION.** The Passivity of Metals, and Its Relation to Problems of Corrosion, U. R. Evans. *Am. Inst. Min. and Met. Engrs.—Tech. Pub.*, no. 205, May 1929, 23 pp., 10 figs. See also abstract in *Iron Trade Rev.*, vol. 84, no. 18, May 2, 1929, pp. 1185 and 1196.

Protective effect of oxide films on metals; experiments in isolation and removal of films from iron; passive and non-passive iron; in latter, protective skin is not sufficiently strong and continuous; rust is considered as not direct oxidation product of metal, but as precipitate formed by secondary reactions; corrosion is caused by differential aeration and not by uniform aeration.

**MACHINABILITY.** Cutting Tools Research Committee—Report on Machinability, E. G. Herbert. *Instn. Mech. Engrs.—Proc. (Lond.)*, no. 4, 1928, pp. 775-801 and (discussion) 802-825, 39 figs.

Investigation limited to machinability of sharp cutting tool used in bar lathe; review of results of private investigations; apparatus provided for test and specifications of test bars used; effect of speed on chip formation; measurement of machinability by hardness of chip; conclusions resulting from investigations are given. Previously indexed from several sources.

**PROPERTIES.** The properties of Metals, O. Lodge. *Engineering (Lond.)*, vol. 127, no. 3304, May 10, 1929, pp. 587-588.

Reference is made to state of mind which made discoveries and state of mind which just missed them; notes on metallic conduction; conductivity of copper-tin alloys. Abstract of lecture before Inst. of Metals.

**SULPHIDATION.** Reduction of the Alkaline-Earth Sulphates in Metallurgical Operations (Sur la réduction des sulfates alcalino-ferreux dans les opérations), G. Charpy and L. Jacqué. *Académie des Sciences—Comptes Rendus (Paris)*, vol. 188, no. 7, Feb. 11, 1929, pp. 473-475.

Authors' experiments indicate that sulphate of calcium and barium and probably of strontium are equally effective for sulphidation of castings, any differences previously observed being due to physical rather than chemical properties of melt; sulphur equivalents of amounts used were 1 to 4 per cent of iron present; reduction of sulphate to sulphide which may be due to carbon or metallic iron is perceptible at 600 and rapid at 900 degrees Cent.

**TESTING.** Tensile Strength and Hardness of Metals, O. Schwarz. *Metallurgist (Supp. to Engineer, Lond.)*, Apr. 26, 1929, p. 64.

Review of paper No. 313 of Forschungsarbeiten auf dem Gebiete des Ingenieurwesens (Berlin) 1929, previously indexed from original source.

**TESTING.** The Mechanical Testing of Materials, R. W. Bailey. *Metropolitan Vickers Gaz. (Manchester)*, vol. 11, no. 192, Mar. 1929, pp. 231-232, 2 figs.

Works testing of metals employed in construction of important high-speed machinery; periscope examination of rotor forging; crystallization during solidification of octagonal ingot. Paper read before Instn. Mech. Engrs.

**X-RAY ANALYSIS.** Studies in Metal Crystal Orientation; Determination of Orientation of Metallic Single-crystal Specimens by High-voltage X-rays, T. A. Wilson. *Am. Inst. Min. and Met. Engrs.—Tech. Pub.*, no. 210, 18 pp., 12 figs.

Discussion of various methods of determining orientation; description of apparatus; Laue photographs, obtained by high-voltage X-ray, may be solved either by application of gnomonic projection alone or by application, of stereographic-gnomonic double net.

#### NICKEL ALLOYS

Developments in Nickel Alloyed Iron, Steel and Cast Iron, J. S. Vanick. *Can. Machy. (Toronto)*, vol. 40, no. 7, Apr. 4, 1929, pp. 42-43 and 81-83.

History of use of nickel in iron and steel; present uses and success of nickel alloys; pure nickel mandrels, jigs, and forgings used in heat-treating rooms, permalloy, invar, elinvar and platinite properties; application of nickel for steel boilers; locomotive forgings; uses of cast steels containing 1 to 3 per cent nickel. Paper read before Am. Soc. Steel Treating.

#### NICKEL-ALUMINUM ALLOYS

Constitution of Nickel-Aluminum Alloys, N. B. Pilling and T. E. Kihlgren. *Am. Soc. Steel Treating—TRANS.*, vol. 15, no. 4, Apr. 1929, pp. 728-729, 1 fig.

Recommended practice committee release:

nickel and aluminum are completely miscible when fused; although compound nickel-aluminum is formed which has melting point nearly 200 degrees Cent. greater than that of nickel; about 14 per cent aluminum is taken in solid solution in nickel, forming alpha phase, while compound nickel-aluminum form solid solutions with as much as 10 per cent nickel, although not with aluminum.

#### NITRIDATION

The Nitride Hardening of Steel. *Metallurgist (Supp. to Engineer, Lond.)*, Apr. 26, 1929, pp. 63-64.

Nitriding consists essentially of heating specimen or part for definite period of time at temperature of 500 degrees Cent. in atmosphere of ammonia; in order to meet different requirements of core strength for different purposes, range of steels, termed nitralloy steels, which contain aluminum, has been developed.

Surface Hardening of Steel by Nitrogen, H. W. McQuaid. *West. Soc. of Engrs.—Jl.*, vol. 34, no. 4, Apr. 1929, pp. 239-248, 2 figs.

Failures and shortcomings as well as successes of nitriding processes are discussed; commercial process of hardening; depth and hardness of nitride case obtained at different temperatures; nitriding prevents corrosion; hardness measured at varying depths; process not yet fully understood.

The Nitriding Process and Its Advantages, J. L. Helquist. *Machy. (N. Y.)*, vol. 35, no. 9, May 1929, p. 647.

Recently developed method of obtaining surface hardness on alloy steel parts, which may supercede ordinary case-hardening for many purposes, is described; molybdenum aluminum steels react most favorably to process.

Working Rules for Nitrogen Hardening, H. W. McQuaid. *Iron Age*, vol. 123, no. 19, May 9, 1929, pp. 1272-1274, 1 fig.

Process of nitrogen hardening with working rules is discussed; high and low-temperature cycle gives deep case and file-hard surface in 18 hours; time of process varies depending on temperature, materials used and future use of article; sealed retorts are used; circulation speeds up work; combination cycle gives best results. Abstract of paper presented before West. Soc. of Engrs.

#### PIPE, CAST IRON

**CENTRIFUGAL CASTING.** Centrifugally Cast Iron Pipes of the L. von Roll Iron Works (Schleudergussrohren der L. von Roll'schen Eisenwerke), M. Ros. *Schweiz. Verband für Materialprüfungen der Technik (Assn. Suisse pour l'Essai des Matériaux)—Bericht (Zurich)*, no. 12, July 1928, pp. 17-36, 23 figs.

Detailed data on comparative tests of centrifugally cast and sand moulded pipes, including bending, tensile, impact and bursting tests; also chemical and microscopical analyses.

#### REFRACTORY MATERIALS

Developments in the Manufacture, Uses and Applications of Refractories, M. C. Booz. *Fuels and Furnaces*, vol. 7, no. 4, Apr. 1929, pp. 567-571.

Development of modern refractory plants; adoption of tunnel kilns for burning fireclay, silica and chrome brick; new products developed; burned magnesite brick, mullite,

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#### SHEET

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#### STEEL

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and sillimanite; refractories, diaspore refractories, light-weight cyanite brick; uses and applications; importance of research being conducted.

#### SHEET STEEL

**CRACKING.** Hair Cracks in the Surface of Steel Plates (Haarrisse auf der Oberfläche von Blechen), A. E. Matejka. *Stahl u. Eisen (Düsseldorf)*, vol. 49, no. 18, May 2, 1929, pp. 643-645, 8 figs.

Notes on prevention of occurrence of hair cracks in surface of plates; careful deoxidation of steel is necessary.

#### SPRING STEEL

Automobile Springs and Materials Employed (Kraftwagenfedern und ihre Werkstoffe), H. Franz. *Motorwagen (Berlin)*, vol. 32, no. 11, Apr. 20, 1929, pp. 224-229, 12 figs.

Factors governing design of springs are enumerated; properties of different spring steels, including unalloyed carbon steel, manganese, silicon, and chromium steels, or steels with two alloy constituents; heat treatment, etc.

#### STEEL

**NITROGEN CONTENT.** Effect of Nitrogen in Iron and Steel, V. N. Svechnikov. *Heat Treating and Forging*, vol. 15, no. 4, Apr. 1929, pp. 436-441, 6 figs.

Review of investigations made on bad effects of nitrogen content on physical properties of steel; sources of nitrogen reviewed; early views regarding nitrogen content; Bessemer steel apt to contain more nitrogen; breakage of bars indicates danger; conditions favoring nitrification; absorption of nitrogen by metaloids. Bibliography.

#### STEEL ANALYSIS

**HYDROGEN DETERMINATION.** A Method for the Estimation of Hydrogen in Steel, T. E. Rooney and G. Barr. *Iron and Steel Inst.—advance paper*, no. 13, May 1929, 8 pp., 4 figs. See also *Engineering (Lond.)*, vol. 127, no. 3304, May 10, 1929, pp. 597-598, 4 figs.

Problem was to determine whether steel which had been in contact with hydrogen under pressure at elevated temperature for considerable time contained appreciable amount of hydrogen; method was adopted of heating sample in known volume of nitrogen and measuring volume of hydrogen evolved by means of Shakespear katharometer; it has been successfully used to estimate hydrogen in samples of mild steel subjected to special conditions.

**OXIDE DETERMINATION.** Determination of Oxygen in Steel According to Hot-Extraction Method (Beitrag zur Bestimmung des Sauerstoffs im Stahl nach dem Heissextraktionsverfahren), G. Thanheiser and C. A. Mueller. *Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung (Düsseldorf)*, vol. 11, no. 5, 1929, pp. 87-94, 5 figs.

Outline of oxide-determination method; review of research on oxide determination according to hot-extraction process; investigation on reduction of manganese oxide, silicic acid and aluminum oxide; influence of manganese evaporation on oxide determination.

#### STEEL FOUNDRY PRACTICE

Casting of Steel in a Water-Cooled Copper

Ingot Mold (Giessen von Stahl in eine wassergekühlte Kupferkokille), W. Ortel. *Stahl und Eisen (Düsseldorf)*, vol. 49, no. 49, May 9, 1929, pp. 696-698, and (discussion) pp. 699-700.

Results of strength and structure investigations and partial determination of hardness zone and magnetic properties of steel cast in water-cooled ingot molds.

#### STEEL INGOTS

**CASTING.** Casting of Steel Ingots (Das Giessen von Stahlblöcken), F. Pacher. *Stahl u. Eisen (Düsseldorf)*, vol. 49, no. 18, May 2, 1929, pp. 627-643, 26 figs.

Significance of casting in comparison with melting process; most important casting processes and their basic differences; thermal metallurgical, and mechanical aspects; temperature conditions with casting and with solidification of different kinds of steel; significance of piping problem and its prevention; segregation and blowhole formation; oxidation during casting.

#### STEEL MANUFACTURE

**RESEARCH.** New Method for the Production of Sound Steel, C. Parsons and H. M. Duncan. *Iron and Steel Inst.—Advance paper (Lond.)*, no. 10, May 1929, 12 pp., 31 figs. See also abstract in *Iron and Coal Trades Rev. (Lond.)*, vol. 118, no. 3192, May 3, 1929, pp. 654-655 and 671, 8 figs.

Method of overcoming segregation and axial unsoundness is based on principle that freezing of ingot should proceed from bottom upwards, isothermals being substantially parallel to bottom of ingot; description of tests and tabular data on results.

#### STEEL METALLOGRAPHY

Studies on Martensite (Estudios sobre la Martensita), J. Orland. *Anales de la Asociación de Ingenieros (Madrid)*, vol. 8, nos. 1 and 3, Jan. and Mar. 1929, pp. 28-39 and 144-153, 21 figs.

Jan.: General discussion of work of investigators, Bibliography. Mar.: Continued discussion and deduction that martensite is not solid solution of gamma iron and carbon, that it is always present in needle-like form and in two distinct phases distinguishable by etching with nitric or picric acid, insoluble white needles being allied with austenite grains.

#### STEEL METALLURGY

On the Stepped Lowering of the Transformation in Steel, T. Murakami. *Tohoku Imperial Univ.—Tech. Reports (Sendai, Japan)*, vol. 8, no. 2, 1929, pp. 245-253, 8 figs.

Note in explanation of eutectoid change in steel; stepped lowering is observed rather than gradual steels when rate of cooling increases; thermal cooling curves, quenching diagram, magnetic and dilatometric cooling curves of various steels and crystallization velocity in relation to temperature are given. (In English).

**MICROSTRUCTURE.** The Transformation of Austenite into Martensite in a 0.8 per cent Carbon Steel, D. Lewis. *Iron and Steel Inst.—Advance paper (Lond.)*, no. 7, May 1929, 11 pp., 17 figs. See also abstract in *Iron and Coal Trades Rev. (Lond.)*, vol. 118,



no. 3192, May 3, 1929, pp. 656-657 and 672, 14 figs.

Tests were conducted on steel quenched in molten tempering salt at varying temperatures, with subsequent slow cooling or rapid cooling. Bibliography.

The Constitution of Steel and Cast Iron Section II, Part VIII, F. T. Sisco. *Am. Soc. Steel Treating—TRANS.*, vol. 15, no. 5, May, 1929, pp. 837-850, 9 figs.

Tempering of hardened steel; object and various operations connected with this phase of heat treatment are discussed; various temperatures used and proper time for holding at tempering temperature; theory of tempering; structural changes taking place in tempering and which occur when high carbon is cooled at various rates followed by tempering; effect of tempering upon structure of high-carbon steel is illustrated by photomicrographs. (Continuation of serial.)

### STEEL PROPERTIES

The Influence of Pickling Operations on the Properties of Steel, H. Sutton. *Iron and Steel Inst.—Advance paper (Lond.)*, May 1929, 24 pp., 10 figs. See also abstract in *Iron and Coal Trades Rev. (Lond.)* vol. 118, no. 3192, May 3, 1929, pp. 674-676, 1 fig.

Description of experiments to determine influence of pickling in acid solutions on several kinds of steel strip; to obtain comparable conditions, all strips used were about 0.015 in. thick and 2.88 in. wide; 10 per cent sulphuric acid was used; other tests were made, using hydrochloric acid, nitric acid, and sulphuric acid containing colloids.

Brittleness in Mild Steel, G. R. Bolsover. *Iron and Steel Inst.—Advance paper (Lond.)*, no. 3, May 1929, 15 pp., 5 figs. See also abstract in *Iron and Coal Trades Rev. (Lond.)*, vol. 118, no. 3192, May 3, 1929, pp. 665-667, 5 figs.

Results of investigation, found to confirm work done on Continent; impact results; effect of temperature; influence of initial condition; effect of phosphorus on blue brittleness; increase of hardness by cold work.

### STEEL RESEARCH

Third Report on the Heterogeneity of Steel Ingots, J. H. Andrew, D. Binnie, C. H. Desch, B. S. Smith and N. M. H. Lightfoot. *Iron and Steel Inst.—Advance paper (Lond.)*, no. 1, May 1929, 71 pp., 40 figs. See also *Iron and Coal Trades Rev. (Lond.)*, vol. 118, no. 3192, May 3, 1929, pp. 640-644, 37 figs.

Liquidus and solidus ranges of some commercial steels, and solubility of iron and manganese sulphides in steel, by Andrew and Binnie; interim report on density of molten steel, by Desch and Smith; part by Lightfoot is described as attempt to determine effect of latent heat on solidification of steel ingots.

### STEEL TEMPERING

Recovery of Austenitic Steels (Sur les revenus d'aciers austénitiques) A. Michel and P. Benazet. *Académie des Sciences—Comptes Rendus (Paris)*, vol. 188, no. 13, Mar. 25, 1929, pp. 912-915.

Special steels which may be rendered austenitic by tempering at high temperatures fall into two classes: there is sudden dilation accompanied by liberation of heat at temperature which depends on composition of steel

and rate of heating; above certain temperature there is gradual negative anomaly in dilatation; the two types of recovery are analogous, and difference is due only to different rates of reaction.

### STRENGTH OF MATERIALS

Rapid Method of Determining Endurance Limit by Means of Measuring Electric Resistance, S. Ikeda. *Tohoku Imperial Univ.—Tech. Reports (Sendai, Japan)*, vol. 8, no. 2, 1929, pp. 167-196, 31 figs.

Increase of electrical resistance measured in steel specimens subjected to repeated bending stress; determination of endurance limit, relation between this limit and Shore's hardness number is traced. (In English.)

### STRUCTURAL STEEL

Steel for Structural Purposes, W. H. Hatfield. *Structural Engr. (Lond.)*, vol. 7, no. 5, May 1929, pp. 173-182 and (discussion) 182-184, 5 figs.

Discussion of microstructure, characteristics and properties of varieties of iron and steel, and accepted factors of safety; corrosion.

CONCRETE ENCASEMENT. A Review of Tests to Determine Effect of Concrete Floors and Fireproofing on Stresses in Structural Steel, R. R. Zippodt. *West. Soc. of Engrs.—Jl.*, vol. 34, no. 5, May 1929, pp. 286-299.

Review of all available literature which has been published on subject, showing that while only few such tests have been made they have been conclusive, that tests on steel encased in concrete show fiber stresses less than calculated stress for open beam, and deflection less than one-quarter of that calculated; tests on beams encased in gunite also show increase in strength. American and English bibliography; 11 references.

SURFACE COMPRESSION. Surface Compression of Steel Structural Parts (Das Druecken der Oberfläche von Bauteilen aus Stahl), O. Foepl. *Stahl und Eisen (Düsseldorf)*, vol. 49, no. 17, Apr. 25, 1929, pp. 575-577, 5 figs.

Results of tests carried out at Woehler Institute, Braunschweig, show that endurance of structural parts can be considerably increased by surface compression; details of bending and torsional tests.

### TOOL STEEL

HEAT TREATMENT. Recommended Practice Committee Releases. *Am. Soc. for Steel Treating—TRANS.*, vol. 15, no. 5, May 1929, pp. 877-880.

Report of tentative general recommendation for heat treatment of tool steels.

### TUNGSTEN CARBIDE CUTTING TOOLS

The present Status of Tungsten Carbide as a Cutting Material, F. C. Spencer. *Am. Soc. Mech. Eng.—Advance paper for mtg.*, May 13 to 16, 1929, 5 pp.

Report of Sub-Committee on Machinability of A. S. M. E. Special Research Committee on Cutting of Metals; brands available and market conditions; some experiences of industry with tungsten carbide; effect on machine tools; summary and recommended action to be taken by committee.



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